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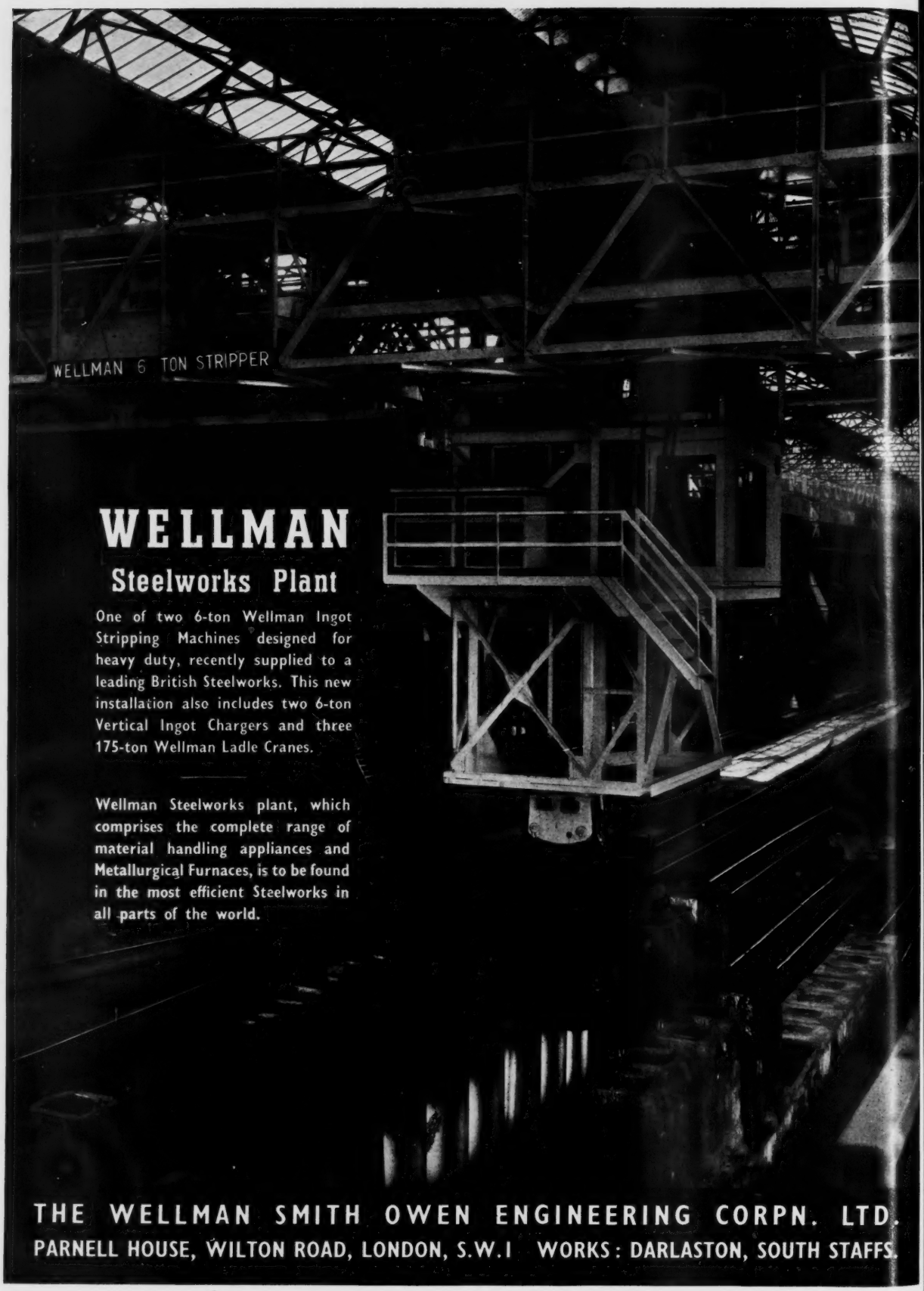

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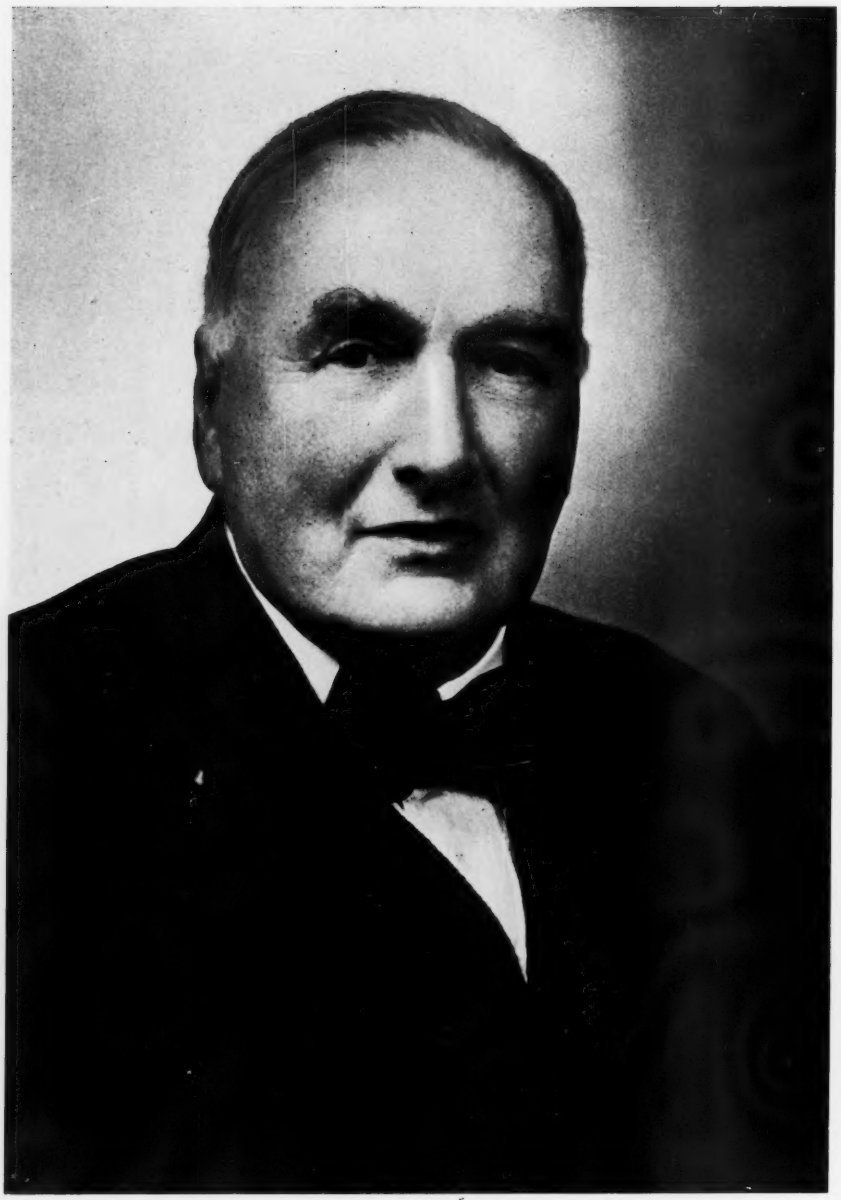
GEORGE BAKER, O.B.E., M.I.Mech.E.

GEORGE BAKER was born at Conisborough in 1878, the third son of the late John Baker, founder of John Baker and Bessemer Limited. He was educated at Rotherham Grammar School and, on leaving school, served an engineering apprenticeship in his father's works in Rotherham. During his apprenticeship he attended evening classes, and he was awarded an Associateship in Metallurgy of Sheffield University in 1899.

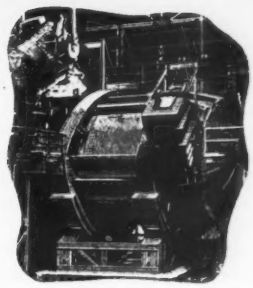
In 1904 Mr. Baker took charge of the construction of the Kilnhurst steel works, and he was appointed Manager of the works on its completion. He became Chairman of John Baker and Bessemer Limited in 1947, a position which he retained until his resignation in 1951. He is at present Managing Director and Chairman of Woodhouse and Rixson Limited, Sheffield.

Mr. Baker was awarded the O.B.E. in 1918, in recognition of his services in the production of munitions in the first World War. He was President of the Rotherham and District Chamber of Commerce from 1931 to 1936, and is a Freeman of the Company of Cutlers in Hallamshire. He became President of the Sheffield Society of Engineers and Metallurgists in 1952, and has been a Member of The Iron and Steel Institute since 1902.



George Baker, O.B.E., M.I.Mech.E.
Honorary Member of Council, 1952-53

Activities of Constituents of Iron and Steelmaking Slags



Part I—IRON OXIDE

By E. T. Turkdogan, Ph.D., M.Met., and J. Pearson, Ph.D., M.Sc., F.R.I.C.

SYNOPSIS

A survey has been made of the effect of temperature and slag composition on the activity of ferrous oxide in molten slags. The activities of the components in the binary 'FeO–SiO_2' system are temperature dependent, but, so far as can be determined, temperature is without effect on the activity of ferrous oxide in more complex slags. It is shown that, for a given iron oxide concentration, small proportions of silica increase the activity of ferrous oxide whereas larger proportions decrease it. Iso-activity curves of ferrous oxide in the system $(CaO+MnO+MgO)$–FeO–$(SiO_2+P_2O_5)$ are drawn on a ternary diagram. Fluoride increases the activity of ferrous oxide in all the slags studied. The effects of composition are discussed on the basis of present knowledge of molten slags.

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Introduction

One of the greatest obstacles to the application of physico-chemical principles to the elucidation of slag–metal and slag–gas reactions is a lack of knowledge of the activities of the reacting species; the use of molal concentrations in expressions of the mass action law frequently leads to highly variable 'constants.' This drawback can be remedied if activities are employed instead of concentrations; the present paper is the first of a proposed series in which it is hoped to evaluate the activities and activity coefficients of slag and metal constituents of interest in iron and steelmaking processes. Since iron oxide in slag participates in numerous reactions between metal, slag, and gas, it was considered that the evaluation of the activity of ferrous oxide should be first studied.

Chipman, Fetters, and Taylor[1, 2] have evaluated the activity of ferrous oxide in slags of a wide range of composition, but it appears desirable to examine their results together with those of other, and particularly more recent, investigators. The experimental results used in this paper are those obtained by Chipman and his co-workers,[1, 2–5] by Körber and Oelsen,[6, 7] and by Schuhmann and Ensio.[8]

Although it is becoming clear from, for example, viscosity, electrical conductivity, and electrolysis measurements that molten slags are ionic in nature and that, in the presence of excess basic oxides, silica, phosphorus pentoxide, etc., are converted to polyhedral oxyacid radicles such as SiO_4'''', PO_4''', etc., no clear evidence is available as to the ionic nature of less basic slags, beyond the fact that the acidic oxides appear to be present as polymeric chains or networks. For this reason and for greater simplicity it has been decided that, in the present discussion, slag compositions should be expressed in terms of mole fractions of oxides, even though these oxides may not exist as such in the melt.

FERROUS OXIDE–SILICA BINARY SYSTEM

The iron oxide–silica system is not strictly a binary system, since there is always some ferric oxide present, particularly under conditions of high oxygen potential. When in equilibrium with iron, however, the proportion of ferric oxide is low and can generally be neglected. The molten slag may thus be regarded simply as a solution of ferrous oxide in silica.

Activity of Ferrous Oxide

The activity of ferrous oxide in a slag can readily be calculated by dividing the oxygen content of liquid iron in equilibrium with it by the oxygen content of iron in contact with wüstite at the same temperature. The latter is given by the equation

$$\log [0\%] = -\frac{6320}{T} + 2{\cdot}734 \quad \ldots\ldots\ldots\ldots (1)$$

This equation will be correct only if the iron, the oxygen content of which is to be used to calculate the activity of ferrous oxide in the slag, contains no alloying elements that raise or lower the activity coefficient of dissolved oxygen. It is known that carbon reduces this activity coefficient below unity,[9] and silicon is reported to have a similar effect,[10, 11] and manganese should behave in the same way. Bookey[12] has shown that dissolved oxygen reduces the activity coefficient of phosphorus, and it is to be expected that phosphorus would in turn reduce the activity of oxygen. Although the metal in equilibrium with the slags under consideration contained small amounts of manganese, silicon, carbon, sulphur, and phosphorus, the effect these would have on the activity coefficient of oxygen is uncertain; it is there-

Paper CP/10/9 of the Chemistry Department of the British Iron and Steel Research Association, received 18th October, 1952. Dr. Pearson is Head of Department and Dr. Turkdogan is in the Physical Chemistry Section of the Chemistry Department of the Association.

fore assumed that the weight percentage of oxygen in iron determined by analysis may be used to represent oxygen activity in iron.

Alternatively, when the slag is in contact with solid iron, the activity of ferrous oxide can be calculated from the composition of the gas phase in equilibrium with it. This was the procedure adopted by Schuhmann and Ensio,[8] who investigated the thermodynamics of ferrous silicates over the temperature range 1260–1360° C. They found that the activity–composition relationships deviated considerably from Raoult's Law but that, within the temperature range studied, the activities of ferrous oxide and silica were independent of temperature. Their results at 1315° C., recalculated on the basis of $N_{FeO} + N_{SiO_2} = 1$, are shown in Fig. 1*a*, together with those calculated by the first-mentioned method from almost pure iron silicate slags studied by earlier workers.

Figure 1*a* shows the variation of activity of ferrous oxide as a function of mole fraction at four different temperatures. In the composition range studied, the system shows a negative deviation from ideality. According to Schuhmann and Ensio, the activity coefficient becomes greater than unity beyond 0·75 mole fraction of FeO, but there is insufficient reliable information to confirm this. It can be seen, however, that the deviation from Raoult's Law decreases as temperature is increased.

The recent results of Fischer and vom Ende[13] indicate that the activity of ferrous oxide in silica-saturated ferrous silicate slags decreases with increasing temperature. The experimental methods of these workers leave much to be desired; they attempted to equilibrate an iron oxide slag with molten iron in a silica crucible, with a surrounding atmosphere of air. It is clear that, even if the slag and metal were in equilibrium, this did not apply to the metal–gas phases. The oxygen contents found in the metal were lower than those obtained by previous workers under similar slags. Assuming that the oxygen analyses of Fischer and vom Ende were correct (and their values for iron under pure iron oxide slags indicate this to be true), then the iron oxide content of their slags was too high. This was probably the result of oxidation of the surface of the metal after the high-frequency heating was switched off and before the surface was covered by slag rising from the body of the melt, and was probably also the result of oxidation of the slag by the air in the furnace. It should be recorded that the ferric oxide content of their slags was much greater than that which would correspond to equilibrium with liquid iron. It is suggested that the slags actually analysed were not the same as those which were in contact with the melt before sampling, and no use has been made of their results in this paper.

In Figure 1*b*, $\log a_{FeO}$ has been plotted against the reciprocal of the absolute temperature for several different ferrous oxide concentrations. Since the variation of activity with temperature at constant composition is given by

$$\log a = \frac{\bar{L}}{4{\cdot}575T} + \text{constant}, \quad \ldots\ldots\ldots\ldots\ldots\ldots (2)$$

the relative partial molar heat content $\bar{L}_{FeO}$ of ferrous oxide can be calculated by multiplying the slopes of the lines by 4·575, on the assumption that $\bar{L}_{FeO}$ is independent of temperature.

Activity of Silica

If the FeO–SiO_2 system is considered to be a true binary system, the activity of silica can be found over a range of composition when the ferrous oxide activity is known. In other words, taking the components of the system as FeO and SiO_2, the activity of the latter can be calculated from the integrated Gibbs–Duhem equation:

$$\log \frac{(a_{SiO_2})}{(N_{SiO_2})} - \log \frac{(a_{SiO_2})'}{(N_{SiO_2})'} = -\int_{N'_{FeO}}^{N_{FeO}} \frac{(N_{FeO})}{(N_{SiO_2})}\, d \log \frac{(a_{FeO})}{(N_{FeO})} \quad \ldots (3)$$

Taking solid silica as the standard state, and $(a_{SiO_2})' = 1{\cdot}0$ when the molten silicate is in equilibrium with solid silica, the second term on the left-hand side of equation (3) becomes a constant, *viz*,

$$-\log \frac{1}{(N_{SiO_2})'},$$

where $(N_{SiO_2})'$ is the silica content of the liquidus for a given temperature.

The variation of silica activity with temperature and composition, calculated from the above equation, is shown in Fig. 1*c*, in which the silica activity is plotted against the mole fraction of ferrous oxide. Each of the isothermals shows a positive deviation

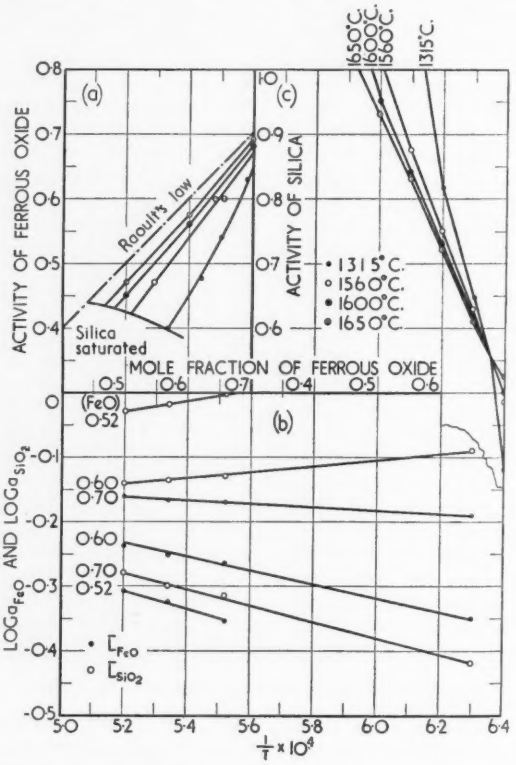

Fig. 1—Influence of composition and temperature on the activities of ferrous oxide and silica in the ferrous oxide–silica system

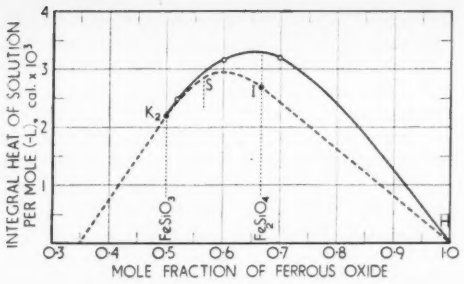

Fig. 2—Heat of mixing in the ferrous oxide–silica system

from ideal behaviour. The deviation, however, becomes slightly smaller as the temperature increases.

Figure 1*b* shows also the effect of temperature on the activity of silica for different ferrous oxide concentrations. The slopes of the lines, multiplied by 4·575, give the relative partial molar heat contents $\overline{L}_{SiO_2}$ of silica (*see* equation (2)).

Heat of Mixing

In considering the heat of mixing in the liquid ferrous oxide–silica system, Rey[14] derived an integral heat of solution curve from the heats of formation and fusion of ferrous oxide, silica, and ferrous ortho- and meta-silicates; part of this curve is reproduced as a broken line in Fig. 2. Rey concluded that, over the range from pure ferrous oxide to silica saturation, ferrous oxide should show an increase in activity with increase in temperature. In the neighbourhood of silica saturation the activity of silica in the melt shows a strong positive deviation. Because the solubility of silica in the melt increases with temperature, the activity of silica in this region decreases with increase in temperature. In the iron oxide-rich region, however, the activity of silica increases with temperature.

Table I

RELATIVE PARTIAL MOLAR AND INTEGRAL HEAT CONTENTS OF PURE FERROUS SILICATE SLAGS

N_{FeO}	$\overline{L}_{FeO}$	$\overline{L}_{SiO_2}$	L
0·52	− 7400	+ 2900	− 2500
0·60	− 5400	+ 350	− 3100
0·70	− 1300	− 7700	− 3200

These conclusions are in good agreement with the present findings illustrated in Fig. 1, and this indicates that the latter could be used to construct the heat-of-mixing curve. The values of $\overline{L}_{FeO}$ and $\overline{L}_{SiO_2}$ calculated from Fig. 1*b* are given in Table I, together with values of the integral heat of solution, $L =$

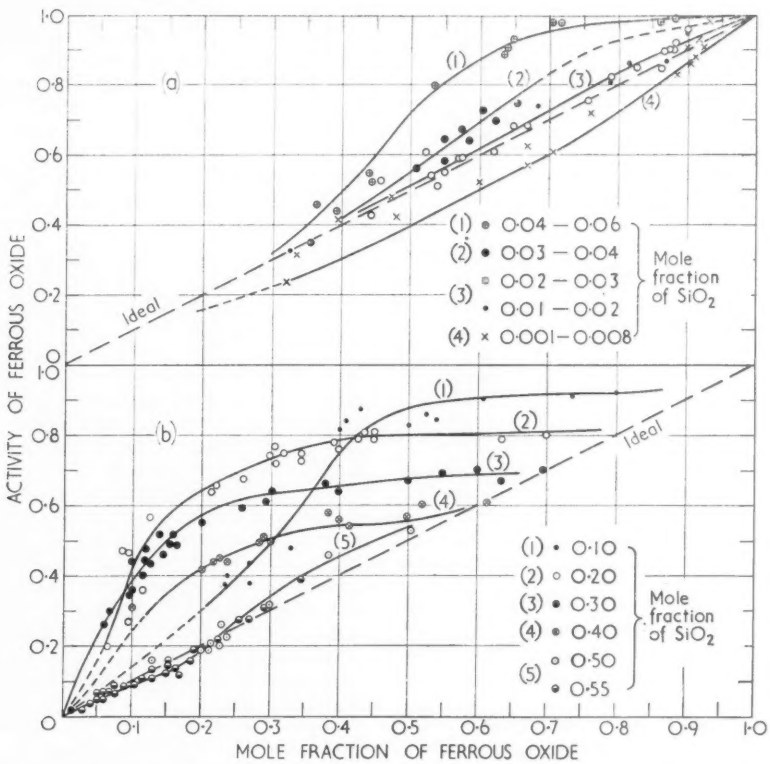

Fig. 3—Influence of composition on the activity offerrous oxide in complex slags

Table II
EFFECT OF TEMPERATURE

N_{FeO}	$N_{SiO_2} + N_{P_2O_5}$	Temp., ° C.	a_{FeO}
0·1	0·2	1550	0·44
,,	,,	1575	0·45
,,	,,	1600	0·44
,,	,,	1625	0·44
,,	0·3	1550	0·40
,,	,,	1575	0·41
,,	,,	1600	0·40
,,	,,	1625	0·39
,,	,,	1650	0·36
,,	0·4	1550	0·25
,,	,,	1600	0·25
,,	,,	1625	0·25
0·2	0·1	1575	0·30
,,	,,	1625	0·30
,,	0·2	1550	0·66
,,	,,	1575	0·66
,,	,,	1600	0·64
,,	,,	1625	0·65
,,	,,	1650	0·66
,,	0·3	1550	0·55
,,	,,	1575	0·56
,,	,,	1600	0·53
,,	,,	1625	0·54
,,	,,	1650	0·54
,,	0·4	1550	0·42
,,	,,	1575	0·43
,,	,,	1600	0·42
,,	,,	1625	0·42
0·3	0·2	1550	0·73
,,	,,	1575	0·71
,,	,,	1600	0·73
,,	,,	1625	0·73
,,	,,	1650	0·71
,,	0·3	1550	0·59
,,	,,	1575	0·60
,,	,,	1600	0·60
,,	,,	1625	0·60
,,	,,	1650	0·59
,,	0·4	1550	0·49
,,	,,	1575	0·50
,,	,,	1600	0·50
,,	,,	1625	0·50
0·5	0·1	1575	0·87
,,	,,	1600	0·88
,,	,,	1625	0·87
,,	0·2	1575	0·79
,,	,,	1600	0·81
,,	,,	1625	0·80
,,	,,	1650	0·82
,,	0·3	1575	0·66
,,	,,	1600	0·66
,,	,,	1625	0·65
,,	,,	1650	0·65
,,	0·4	1550	0·59
,,	,,	1575	0·58
,,	,,	1600	0·60
,,	,,	1625	0·58
,,	,,	1650	0·57

$N_{FeO}\overline{L}_{FeO} + N_{SiO_2}\overline{L}_{SiO_2}$. To make the results comparable with those of Rey, the values of $\overline{L}_{SiO_2}$ have been corrected for the heat of fusion of silica (2000 cal./g. mole) because of the change of standard state from solid to liquid silica.

The full line in Fig. 2 is drawn through the values of L given in Table I. This confirms the correctness of the position chosen by Rey for point K_2 but shifts the maximum nearer to the orthosilicate composition. A tangent at the point S (silica saturation at 1350° C.) to either curve indicates a low value for the heat of solution of liquid silica, a result not in disagreement with Schuhmann and Ensio's findings of a zero heat of solution.

The proposed curve shows that the heat of mixing at the orthosilicate composition is — 3300 cal., about 600 cal. higher than that used by Rey. This suggests that the heat of fusion of the orthosilicate is 15,900 cal. rather than the 17,700 cal. which Rey used. The former value corresponds to an entropy of fusion of 10·8 cal./° C. per mol., a not unreasonable figure.

Although the results considered in this paper are not adequate to permit calculations of the heats of mixing at mole fractions of ferrous oxide greater than 0·7, they do appear to indicate that the portion of the heat of solution curve beyond this concentration is curved. If the activities in this composition range are required, they can be calculated from Fig. 2 and equation (2), the value of the constant for each composition being derived from the results of Schuhmann and Ensio.

(CaO + MnO + MgO)–FeO SYSTEM

It has not been possible to select data on the pure individual CaO–FeO, MnO–FeO, and MgO–FeO systems in equilibrium with molten iron, because the slags so far studied have contained mixed bases in varying proportions, as well as silica. Although some of them contained very little silica and could thus be regarded as silica-free, it was not possible to treat the bases separately.

Curve (4) in Fig. 3*a* shows the relationship between the activity of ferrous oxide and its concentration in slags containing varying proportions of CaO, MnO, MgO, and not more than 0·008 mole fraction of silica. This curve, which is drawn slightly lower than the location of the points would justify, represents approximately the variation of activity of ferrous oxide in silica-free slags (*see* p. 222). The system shows a small negative deviation from Raoult's law.

(CaO + MnO + MgO)–FeO–(SiO_2 + P_2O_5) SYSTEM

The possible variations in composition in this system cover essentially the entire range of basic and acid steelmaking slags.

Effect of Temperature

The activities of ferrous oxide for various slag compositions and temperatures have been calculated and are given in Table II, which groups slags of constant N_{FeO} and ($N_{SiO_2} + N_{P_2O_5}$), the values of $N_{P_2O_5}$ being in all instances less than 0·1. The values of a_{FeO} are averages of results from slags having these characteristics. It can be seen that, at constant ' composition,' temperature is without effect on the activity of ferrous oxide. Fetters and Chipman[1] also noted this and reported that the activity of ferrous oxide "as a function of mole fractions in the quasi-ternary system FeO_f–SiO_2–(CaO + MgO)" is applicable to any temperature within the range of their experiments.

Effects of Composition

The slags have been considered as belonging to a pseudo-ternary system, (CaO + MnO + MgO)–FeO

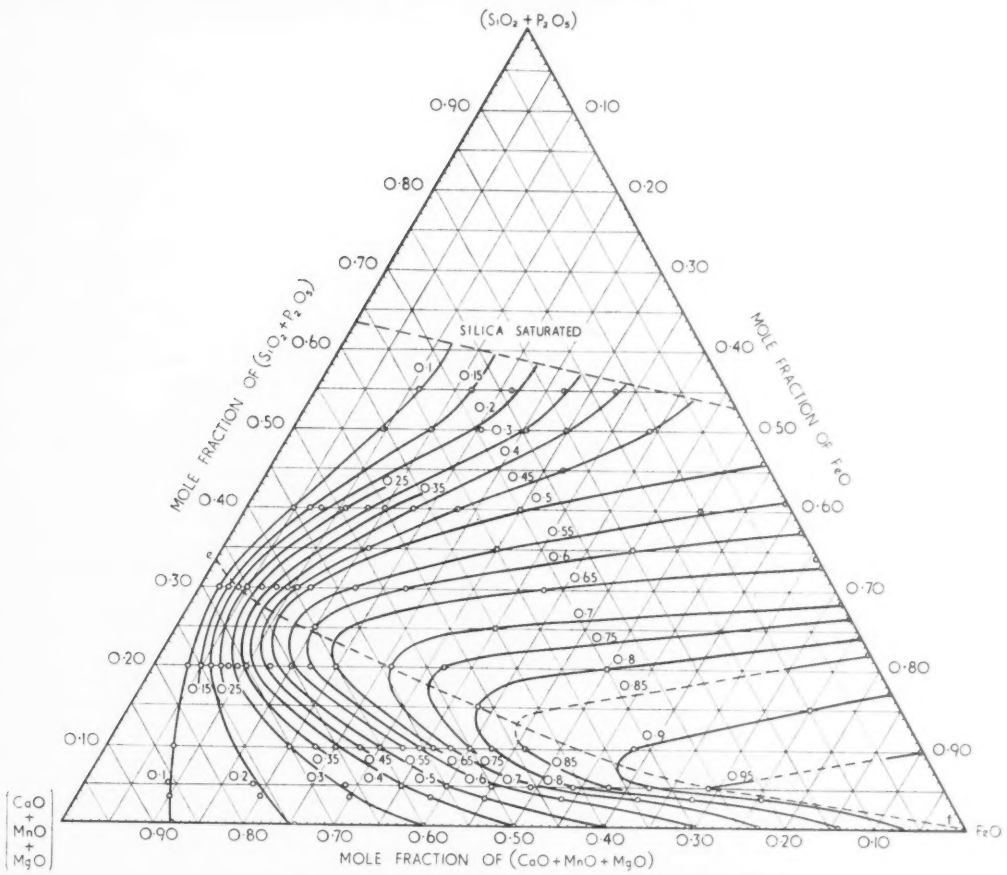

Fig. 4—Activity of ferrous oxide in complex slags

$-(SiO_2 + P_2O_5)$. The effects on ferrous oxide activity of silica and phosphorus pentoxide appear to be additive, at least up to 0·1 mole fraction of the latter.

The variation of activity with change in concentration at several essentially constant silica concentrations is shown in Fig. 3*a*. Silica, even in very small proportions, exerts a marked influence on the activity of ferrous oxide; the positive deviation becomes progressively greater as the silica concentration increases up to 0·06 mole fraction. Figure 3*b* shows the effect of high concentrations of silica. In each group the silica content varies by approximately 0·02 mole fraction about the stated mean. These diagrams apply for any temperature.

It is of interest that the activity of ferrous oxide is determined by the concentrations of silica and ferrous oxide, despite the fact that the relative proportions of CaO, MnO, and MgO may vary widely. In other words, so far as their effect on ferrous oxide activity is concerned, these three bases are interchangeable. There may, of course, be minor differences in their effects, but the existing data do not appear to reveal any.

At a silica concentration of 0·5–0·6 mole fraction, the activity coefficient of ferrous oxide is slightly less than unity.

By making use of the curves in Figs. 1*a* and 3, it is possible to draw iso-activity curves for ferrous oxide in the pseudo-ternary system. This has been done in Fig. 4, which (except possibly for the $FeO–SiO_2$ binary side) holds for any temperature at which the slag is liquid. Along the $FeO–SiO_2$ side the results are those for 1600° C., and those for the FeO-rich end of this side have been calculated from the heat of mixing curve (Fig. 2) by the method mentioned on p. 219.

The inflexion points of the iso-activity curves lie on a smooth curve '*ef*,' where the point '*e*' corresponds to a 2 : 1 base: silica ratio, and '*f*' is the iron oxide corner.

Effect of Fluoride

The effect of fluoride on slag–metal equilibria cannot yet be regarded as satisfactorily elucidated, although the addition of fluorspar to basic steelmaking slags has long been recognized as beneficial. It has been thought that fluorspar increases the fluidity of the slags, which can then dissolve more lime.

Oelsen and Maetz[15] have investigated the influence of fluorspar on the phase relationships of silicate and phosphate systems, but there are only two known sources[4, 5] from which the effect of fluoride on slag–

metal reactions may be deduced. The activity of ferrous oxide in slags containing fluoride, calculated from these two sources, is plotted in Fig. 5 against mole fraction of fluorine for certain concentrations of silica and iron oxide. For fixed FeO and SiO_2 concentrations, the activity of ferrous oxide increases regularly with the fluoride content of the slag. The same behaviour persists even at very low iron oxide activities, as is shown by the inset diagram in Fig. 5.

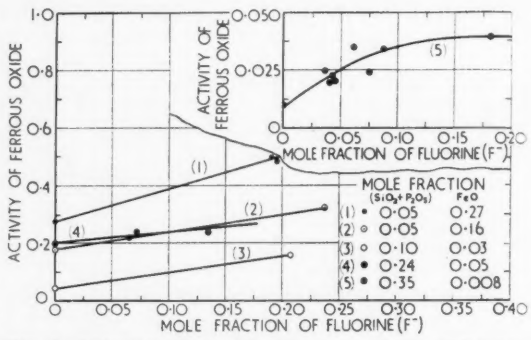

Fig. 5—Influence of fluoride on the activity of ferrous oxide in slags

DISCUSSION

The influence of temperature and slag composition on the activity of ferrous oxide has already been discussed, and it is felt that an attempt should be made to account for some of the observed characteristics.

Temperature has no appreciable effect, except in the simple FeO–SiO_2 system. In this case, Rey[14] has shown by computations from heats of formation and fusion that, except at one composition, the heat of mixing is not zero (*see* Fig. 2). It follows from this that over some part, if not all, of the composition range the relative partial molar heat content of ferrous oxide is also not zero. It then follows from equation (2) that the activity of ferrous oxide must vary with temperature.

In discussing the effect of composition on the activity of ferrous oxide, it is advantageous to treat it in three ways—at constant concentration of FeO, at constant ($SiO_2 + P_2O_5$), and at constant (CaO + MnO + MgO). The activities of ferrous oxide across sections of the ternary diagram (Fig. 4) at 0·3 mole fraction of FeO, ($SiO_2 + P_2O_5$), and (CaO + MnO + MgO), respectively, are given in Fig. 6. It should be noted that, in deriving the various diagrams in this paper, all iron oxide in the slags has been assumed to be ferrous. The actual amounts of ferric oxide present were insufficient to cause any essential errors in the calculations of ferrous oxide concentrations, but their presence must be taken into consideration when discussing activity of ferrous oxide.

Figure 6*a* shows that, at a fixed concentration of FeO, the activity coefficient of this component is less than unity at the base-rich end. In the absence of silica and phosphorus pentoxide, the ferric oxide content of the slag is likely to be present as complex ferrite ions such as Fe_2O_4'', Fe_2O_5'''', FeO_3''', etc. Whatever the nature of these ferrite radicals, their formation implies a reduction in the activity of the ferric oxide. Because of the existence of an equilibrium between ferric oxide, ferrous oxide, oxygen, and iron, this reduction in activity of ferric oxide will result in a reduction in the activity of the ferrous oxide. In the absence of bases, ferrous oxide will tend to form ferrous silicates, which will also cause a reduction in the activity of the ferrous oxide.

Additions of silica or phosphorus pentoxide to a ferrous oxide-base melt should raise the activity of the ferrous oxide, for the following reasons. Electrostatic considerations predict that the stability of trivalent iron complexes should be low because Fe^{+++} has a larger ionic radius and a smaller electric charge than silicon or phosphorus have in their complex ions. Small additions of silica or phosphorus pentoxide will therefore remove oxygen ions from the ferrites and so increase the activity of the ferric oxide and in turn that of ferrous oxide.

Similarly, because the alkaline earth silicates and phosphates are more stable than the ferrous compounds, small additions of bases to a ferrous silicate slag will form calcium, etc., silicates and free the ferrous oxide; this will result in a corresponding increase in activity.

The positive deviation may be ascribed to the fact that the internal pressures of the (CaO + MnO + MgO)–FeO and FeO–SiO_2 systems are different and thus give rise to a positive deviation on mixing[16]; or, alternatively, to the fact that bonds formed between the bases and silica on mixing the two systems weaken the bonds between the ferrous oxide and the bases on the one hand and between the ferrous oxide and the silica on the other. The antagonism between external and internal bonds has been discussed by Frenkel.[17] The positive deviation therefore increases as the proportion of silica increases or as the proportion of bases increases, and tends to a maximum at some particular composition. No special significance is attributed to the occurrence of this maximum at 0·19 mole fraction of SiO_2 in Fig. 6*a*; it occurs at a different concentration of SiO_2 for other proportions of FeO. The result is that, in practically the whole range of composition over this section, the deviation from Raoult's Law is positive.

Herasymenko[18] has shown, by considering basic slags to be fully ionized, that the activity of ferrous oxide increases as the basicity decreases, *i.e.*, as the relative proportion of silica increases. He attributed this to an increase in the activity of the ferrous ion, claiming that the oxygen ion behaved ideally. This was an unjustifiable assumption based on the approximate constancy of the sulphur–oxygen equilibrium constant. However, the result, not unexpectedly, is in agreement with the present findings over the range up to 0·19 mole fraction of SiO_2 in Fig. 6*a*.

Figure 6*b* shows how the activity of ferrous oxide changes over an iso-silica section, 0·3 mole fraction of ($SiO_2 + P_2O_5$). As the relative proportion of ferrous oxide increases, its activity does so also, but not uniformly; there is first a steep rise and the curve tends to flatten out. The rapid rise may be attributed to the fact that, whereas at very low FeO concentrations there is base in excess of the silica and this excess is able to produce a reduction in activity of

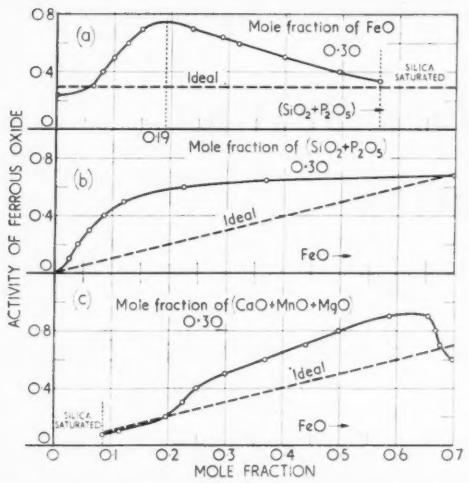

Fig. 6—Influence of slag composition on the activity of ferrous oxide

ferrous oxide (as described above), at higher FeO contents there is no excess base to produce this reduction. The slower rise that follows may be attributed to the fact that, although the FeO content is rising, increasing proportions are being involved in silicate formation.

Figure 6*c* shows results for an iso-bases section, 0·3 mole fraction of (CaO + MnO + MgO). Similar considerations apply in this case, the rapid decrease in activity at high ferrous-oxide concentration being attributable to increased ferrite formation.

It should be noted that over the whole ternary system, except in the vicinity of the FeO–SiO_2 and (CaO + MnO + MgO)–FeO sides, the deviations from ideality are positive.

In considering the effects of fluoride additions to basic slags, Herasymenko[18] stated that fluoride ions increase the activity coefficient of ferrous ions. As already pointed out, this really means that ferrous oxide activity is increased, as has been shown in Fig. 5. It is suggested that the fluorine ion exerts this effect by entering into the silicate and phosphate ions and so liberating oxygen ions. This will lead to an increased activity of ferrous oxide. Experimental work on the influence of fluoride is to be undertaken.

CONCLUSIONS

Present knowledge of the constitution of molten slags does not allow satisfactory representation of composition in terms of ions, except perhaps in basic slags. The use of ion concentrations derived by empirical methods does not appear to have any advantage over the use of mole fractions of the slag components, considered as oxides; mole fractions of oxides were therefore employed in this paper.

The activity of ferrous oxide in any slag has been calculated by dividing the oxygen content of the iron in equilibrium with it by the oxygen content of iron in equilibrium with pure wüstite, at the same temperature.

The activities of ferrous oxide and silica in the FeO–SiO_2 binary system have been shown to be dependent on temperature. A similar conclusion is reached from considerations of heats of formation and fusion. Over the major part of the system, deviations from ideality are negative and this is attributed to silicate formation.

In slags free from silicate and phosphate, the activity coefficient of ferrous oxide is less than unity, as a result of the formation of ferrites.

In slags covering the compositions of basic and acid steelmaking practice, the iron oxide activity does not appear to be affected by temperature between the limits 1550° and 1650° C. In general, the ferrous oxide activity shows positive deviation from ideality.

Iso-activity curves drawn in the ternary system (CaO + MnO + MgO)–FeO–(SiO_2 + P_2O_5) are located somewhat differently from those of Taylor and Chipman,[2] and the locus of the inflexion points of these iso-activity curves is not a straight line, as suggested by them, but a curve which cuts the (CaO + MnO + MgO)–(SiO_2 + P_2O_5) side at a point corresponding approximately to a 2 : 1 base : acid ratio.

The activity of ferrous oxide has been found to be a function of the fluoride content of the slag.

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XUM

Effects of High-Temperature Heating on the Isothermal Formation of Bainite

By

S. A. Cottrell, Ph.D., and T. Ko, Ph.D.

SYNOPSIS

The effects of high-temperature heating on the rate of the austenite–bainite transformation in En 17, 18, 19, 21, 25, and 31 steels have been investigated. The incubation period is shortened and the rate of transformation is increased in Ni–Cr–Mo steels after austenitizing at high temperature. The effect can be removed by re-austenitizing at a lower temperature. These effects were not found in the other steels. 731

HIGH austenitizing temperatures have rarely been used when studying the isothermal decomposition of austenite, except to improve homogeneity. An increase in the austenitizing temperature is known to retard the start of decomposition to pearlite, but that to bainite is unaffected by heating the austenite to 1100° C.[1] The effects of higher temperatures have received little attention. Preece and his co-workers[2] reported that heating above the overheating temperature of a steel had little effect on the position of the isothermal transformation (I.T.) diagram after re-austenitizing, but that the distribution of the pre-eutectoid ferrite had been affected. Hultgren[3] found that, in a series of Mn steels, austenitization at 1300° C. considerably retarded the decomposition of austenite to bainite, in comparison with the decomposition after treatment at lower austenitizing temperatures. However, Winterton[4] observed that, in a Ni–Cr–Mo steel (En 28), austenite quenched from 1200° C. decomposed to bainite more rapidly than that quenched from 850° C.

Investigating the effect of high-temperature heating followed by air-cooling on the tensile properties of six Ni–Cr–Mo steels, Wheeler, Kondic, and Ko[5] found that heating these steels to a certain range of temperature between 1200° and 1450° C. caused the appearance of bainite in the air-cooled specimens. Specimens cooled at the same rate were martensitic after austenitizing at 850° C. The change in structure was accompanied by an increase in ductility and a decrease in tensile strength.

They suggested that heating to high temperatures accelerated the bainite reaction. In these specimens continuous-cooling transformation diagrams in the

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Table I

ANALYSES OF STEELS USED

(Principal alloying elements are shown in italics)

Steel En	Analysis, %								Steelmaking Process
	C	Ni	Mn	Cr	Mo	Si	S	P	
17	0·37	...	*1·51*	...	*0·45*	0·17	0·035	0·027	...
18	0·48	0·18	0·86	*0·98*	0·04	0·25	0·021	0·023	Acid O.H.
19	0·41	...	0·64	*1·37*	*0·28*	0·25	0·030	0·027	...
21	0·33	*3·47*	0·74	0·07	...	0·23	0·027	0·031	Acid O.H.
25 (VA)	0·33	*2·59*	0·68	*0·53*	*0·58*	0·20	0·029	0·029	Acid O.H.
25 (VB)	0·32	*2·60*	0·63	*0·51*	*0·55*	0·22	0·023	0·023	Basic elec.
26	0·42	*2·53*	0·67	*0·72*	*0·48*	0·31	0·022	0·029	Acid O.H.
31*	1·08	0·33	0·53	*1·46*	0·06	0·25	0·015	0·022	Elec. arc

* Containing also 0·11% W

bainite range were intersected by the cooling curve, after the reaction had been sufficiently accelerated by high-temperature heating. The amount of bainite formed thus depended on the extent of activation and, with a given cooling rate, varied with the hardenability of the steel.

If this explanation is correct, a similar effect on the isothermal decomposition of austenite in the bainite range could be expected. This has been investigated.

EXPERIMENTAL METHODS

Details of the steels used are given in Table I. Steels En 18, 21, and 31 were from the stock of steels used for determining I.T. diagrams given in the "Atlas of Isothermal Transformation Diagrams of B.S. En Steels",[6] and were received in the form of 0·116-in. dia. wire. Steels En 17, 19, and 25 (VA and VB) were received as 0·5-in. dia. annealed bars. The En 25 steels, which had been used in previous investigations,[5, 7, 8] were examined both in the as-received condition and after cold-drawing to 0·116-in. dia. wire, with several intermediate annealings. No difference was observed in the incubation periods for intermediate transformation for the two conditions. A few experiments were made with steel En 26, which was from the stock used in determining the I.T. diagram in the Atlas.[6]

Microscopic observations and dilatometric measurements were made.

For metallographic investigations two sizes of specimen were used. The larger size, 0·25 in. long and 0·25 in. dia., machined from the centre portion of the bars, was used for En 17, 19, and 25. The smaller specimens, 0·25 in. long and 0·110 in. dia., were used for En 18, 21, 25, and 31.

The specimens were heated to 950°, 1250°, or 1350° C., in a vertical furnace,[7] under an argon atmosphere. A nitrogen atmosphere gave the same results in En 25 steels. At the end of each period of austenitization the specimens were dropped into a salt bath for the isothermal treatment and were withdrawn after the required periods and quenched into brine.

For dilatometric investigations a nickel-plated specimen, 1·0 in. long and 0·110 in. dia., was placed in a silica dilatometer, heated in a platinum-wound furnace in a stream of nitrogen, and quenched into an isothermal lead or lead–tin bath, unless otherwise stated.

The austenitizing temperatures were controlled to within ± 5° C., the isothermal baths to within ± 2° C. The time required to cool the specimen from 950° C. to a temperature 5° above the I.T. temperatures was 12–15 sec. in the dilatometric method and 18–20 and 25–28 sec. for 0·110-in. and 0·25-in. dia. specimens, respectively, in the microscopic method.

RESULTS

Effect of High Austenitizing Temperature

The times required for the isothermal formation of 3% of bainite in En 25 and En 31, after austenitization at 950° C. for 60 min., at 1250° C. for 15 min., or at 1350° C. for 15 min., are shown in Fig. 1. The diagram for En 31 includes the curves in the Atlas[6] for the same steel. The extra curves in the diagrams for VA and VB steels are for En 25 steel of very similar composition, from the Atlas.[6]

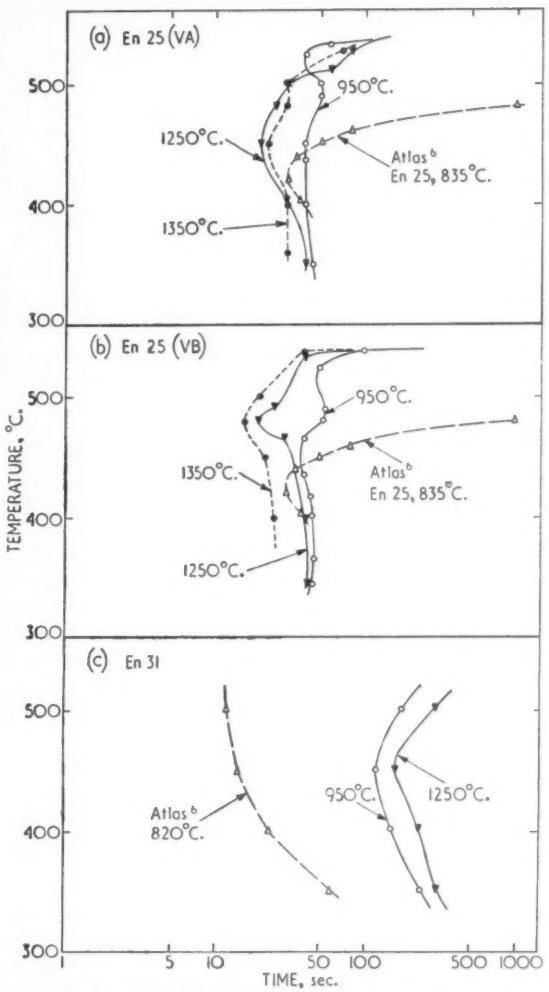

Fig. 1—3% Isothermal transformation curves for steels En 25 (VA), En 25 (VB), and En 31 after austenitizing at the temperatures indicated

The present investigations show that, in both VA and VB steels, bainite started to form earlier after heating to higher temperatures. The effect of high austenitizing temperatures on the rate of transformation was even greater in these two Ni–Cr–Mo steels and in the En 26 steel. This can be seen from the dilatation/time curves in Fig. 2. It appears also that, at temperatures high in the bainite range, the total dilatation was greater, more austenite having transformed, after austenitizing at high temperatures.

The other steels, except En 31, were not noticeably affected by the high-temperature treatment. There was a slight increase in the incubation period and in the rate of transformation, but the effect was too small to be convincing. In En 31 the time required to form 3% of bainite was longer (Fig. 1*c*), and the rate of transformation was lower (Fig. 3), when the

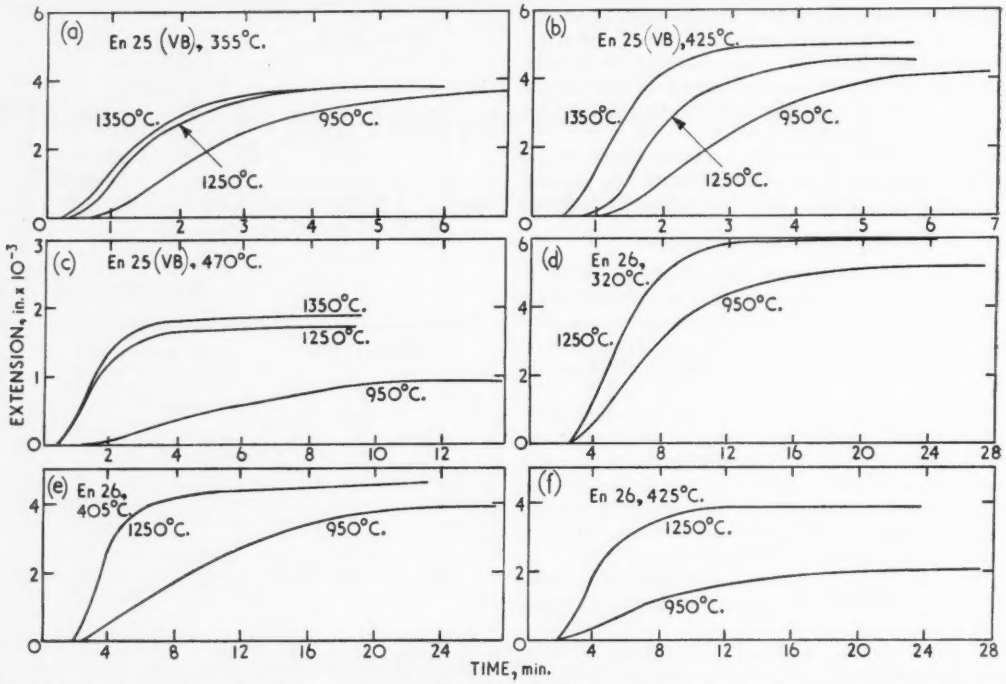

Fig. 2—Dilatation/time curves for steels En 25 (VB) and En 26, quenched from temperatures indicated on the curves in a bath at 320°, 355°, 405°, 425°, or 470° C.

austenitizing temperature was raised from 950° to 1250° C. The curve for this steel in the Atlas[6] (*see* Fig. 1*c*) was obtained after austenitizing at 820° C. and represents the decomposition of austenite with lower chromium and carbon contents, owing to the presence of residual carbide.

Removal of the Accelerating Effect

The accelerating effect of high austenitizing temperature on bainite formation would obviously be of great importance if it persisted through later heat-treatment cycles. To determine whether or not the effect is permanent, specimens of En 25 (VA and VB) and En 26 steels were quenched to room temperature or isothermally decomposed in the bainite range after austenitizing at 1250° C. Their isothermal transformation at 375° and 435° C. was then studied after re-austenitizing at 950° C. In both series of heat-treatments the effect of the previous high-temperature heating was completely removed. There was, in fact, a very slight retardation compared with a specimen which had not been heated above 950° C., probably due to the homogenization at 1250° C. This would remove regions of low alloying content and increase the time required for the start of the transformation.

These experiments also provided ample evidence that the accelerating effect was not due to any decarburization or to any change in chemical composition of the steel.

When the steels were cooled directly from 1250° to 950° C. and held there for various periods up to 16 hr., a partial reversion occurred so that approximately half the accelerating effect remained, as can be seen from Fig. 4.

Microstructure

Several features of the bainite formed after different heat-treatments were common to all steels, with the possible exception of En 31.

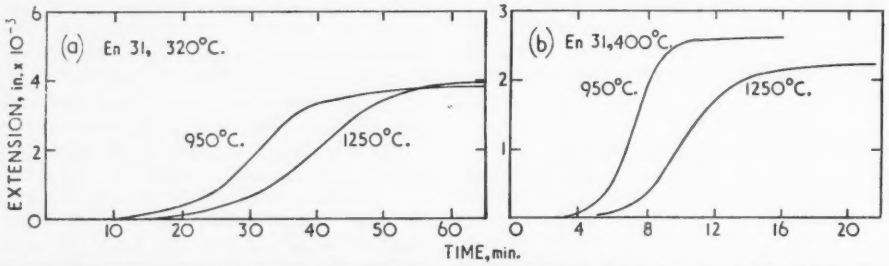

Fig. 3—Dilatation/time curves for steel En 31, quenched from temperatures indicated on the curves in a bath at 320° or 400° C.

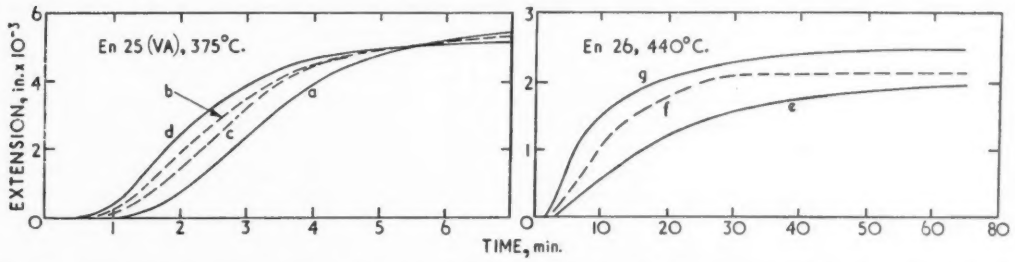

Fig. 4 (*a–d*)—Dilatation/time curves for steel En 25 (VA) quenched in a bath at 375° C.: (*a*) from 950° C.; (*b*) and (*c*) from 950° C. after cooling from 1250° C. and holding at 950° C. for (*b*) 30 min. and (*c*) 120 min.; (*d*) from 1250° C.;

(*e–g*)—Dilation/time curves for steel En 26 quenched in a bath at 440° C.: (*e*) from 950° C.; (*f*) from 950° C. after cooling from 1250° C. and holding at 950° C. for 16 hr.; (*g*) from 1250° C.

The high-temperature austenitization produced, in general, coarser bainitic structures, undoubtedly owing to the increasing austenite grain size. Examination of the structure in the early stages of transformation indicated that the coarse structure was the result of less restricted growth.

There was a greater tendency for bainite to assume an acicular appearance when the steel had been austenitized at 1250° or 1350° C. This was particularly evident in En 31. After austenitizing at 950° C. the bainite tended to be irregular in shape, forming islands in the matrix, although each often consisted of more than one needle. After austenitizing at 1250° C. the bainite was much more acicular.

Figure 5 shows the structure of bainite in the early stages of transformation in En 19, another Cr steel. The acicular nature of the products is evident, but as the transformation progressed, the acicularity became less prominent, owing to crowding. This photograph should be compared with that given in the Atlas,[6] page 30, for a similar steel isothermally decomposed at the same temperature, 400° C.

In several steels, particularly En 17, 19 and 21, austenitizing at 1350° C. promoted the formation of 'feathery' clusters of bainite needles growing from the austenite grain boundaries (Fig. 6).

In the low and medium carbon steels the bainite formed at temperatures high in the range was light-etching (with nital), giving the appearance of outlined ferrite, usually with speckling within the structure Fig. 7); but the bainite formed above 530° C. in En 21 is generally free from carbide particles.[9] In En 19, 21, and 25 the transformation products formed at the same temperature in the early stages seemed to contain slightly more numerous carbide particles after austenitizing at high temperatures, and therefore appeared darker-etching (with nital) at low magnifications (Figs. 8–11). Figures 9 and 10 also show thickening of the bainite plates.

In Cr steels the carbide particles are more numerous, and probably finer, than in other steels; the bainite of En 18, 19, and 31 appeared dark-etching (with nital), even up to 450–500° C. The precipitation within the needle shown in Fig. 11 (En 31 steel) indicates a gradual transition of precipitation from the centre of the needle to its extremity. Precipitation and coalescence were therefore more advanced in the centre of the plate. These observations are consistent with the authors' views on bainite formation.[8]

DISCUSSION

Microstructure

Bainite forms by a process of nucleation and slow coherent growth.[8] It has been suggested that the coherent growth of bainite is controlled by the removal of carbon, either by diffusion into the austenite or by precipitation as carbide within the bainite.

The effects of high austenitizing temperature on the microstructure can be accounted for by the presence of unevenly distributed alloying elements in steels austenitized at low temperatures. Submicroscopical heterogeneities will hinder the full growth of acicular bainite. This suggestion is consistent with the observation that bainite formed in En 31 after austenitization at low temperatures was least aciform. The increasing acicularity and grain size after high-temperature austenitization lead to the more frequent observation of parallel plates originating from the grain boundaries. Bainite formed in the upper range resembles low-carbon martensite by forming needles parallel to one another,[7] whilst lower bainite and high-carbon martensite are formed along more numerous habit planes.

When the distribution of carbide-forming elements in the austenite is not uniform, carbon will tend to segregate. The regions low in carbon and in alloying elements will tend to transform first. When the products formed at the same temperature, in a given time, with and without high-temperature austenitization, are compared, that formed from the low-temperature austenite in the early stages will tend to contain less carbide, and will appear light-etching (with nital).

The removal of regions of low alloying content will retard the beginning of the transformation. It will also reduce the rate of nucleation in the earlier stages of the transformation. On the other hand, it may reduce the restriction to growth. The overall effect on the rate of transformation is, therefore, not certain, because a greater number of potential nuclei would be annihilated by the growing bainite structure. Quantitative measurements of rates of nucleation and growth must be carried out before more detailed examination is possible. In En 31 the effect of homo-

genization on nucleation is probably greater, and the transformation is slower, after high-temperature heating.

Accelerating Effect of High-Temperature Heating

The investigation also shows that the transformation of austenite to bainite in Ni–Cr–Mo steels is accelerated as a result of high-temperature austenitization. The previous investigations showed that the strength of these steels, after air-quenching, was affected by high-temperature heating.[5] There was a fairly continuous decrease in the tensile strength of VA steel, after air-quenching, as the austenitizing temperature was increased from 950° to 1400° C. The maximum strength of VB steel decreased steadily up to 1150° C., above which it remained fairly constant up to 1400° C. The 3% transformation curves (Fig. 1) do not indicate any such difference. The dilatation/time curves (Figs. 2–4) show, however, that VA steel was more affected by the high-temperature heating in the lower part of the bainite range, whereas VB steel was accelerated more in the upper range. Thus, on continuous cooling, the proportion of the transformation products formed at various temperatures will be different in the two steels. This explanation, although reasonable, should not be taken as conclusive, as the kinetics of growth of bainite nucleated at a higher temperature during continuous cooling are at present completely unknown.

No satisfactory explanation can be advanced for the effect of high austenitizing temperature on the transformation characteristics of Ni–Cr–Mo steels in the bainite range. It is unlikely that the austenitic grain size or any fundamental aspect of the mechanism of bainite formation is involved, because (i) the phenomenon is not general and there was no evident difference in the grain size among various steels, as shown by the following figures:

Steel	Austenitized at: 950° C., 1 hr.	1250° C., 15 min.	1350° C., 15 min.
	A.S.T.M Grain-Size Number		
En 17	5–6	...	1–2
En 18	6	4	2
En 19	5	...	2
En 21	5	3	2
En 25 (VA)	5	2	1
En 25 (VB)	6	2–3	1
En 31	5	2	...

and (ii) the effect can be reduced by annealing at 950° C. On the other hand, the effect cannot be due to any changes in composition of the steel as a whole, because the effect vanishes after decomposition of the austenite.

One plausible explanation is that the effect of high-temperature heating is the result of redistribution of non-metallic inclusions. Heating to high temperatures will increase the solubility of sulphur in austenite. On cooling, these elements will reprecipitate as numerous fine sulphide particles.[5, 7, 10] The lattices of these particles are probably crystallographically matched to the austenite lattice at the temperature of precipitation, as they have been found, when visible, to precipitate in a Widmanstätten pattern.[5,7] Then, on cooling to below the temperature of precipitation, the austenite lattice would become strained by these particles and the nucleation of bainite would be easier. The incubation period is shortened, and the rate of transformation to bainite will be increased, if the opposite effect of homogenization on nucleation is slight. It will be relatively small, also, in steels in which transformation is fast, such as En 21. Long-time annealing at 950° C. may cause some coalescence of the precipitate, but most of the straining effect after cooling would remain, and the effect cannot be completely removed. Reheating the steel to 950° C., on the other hand, generates a new set of austenite grains of different orientations. As a result, the straining effect of the inclusions due to the enforced crystallographic matching of lattices would be relieved and the accelerating effect of the high-temperature heating would be completely removed.

Acknowledgments

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LARGE-SCALE TEST COKING PLANT

The British Coking Industry Association has decided to erect a large-scale test coking plant for carrying out experiments on a wide range of coals for the production of coke for blast-furnaces and for domestic and industrial purposes.

The plant is to be designed and operated by the British Coke Research Association, and will include five continuously operating coke ovens; coal reception, blending and charge preparation equipment, including service bunkers; coke-handling equipment; and simple by-product recovery plant. Full instrumentation will be provided to maintain an accuracy of measurement higher than in normal plant practice.

It is considered that much of the research work already carried out by the Association needs full-scale operation to ensure proper correlation with normal practice.

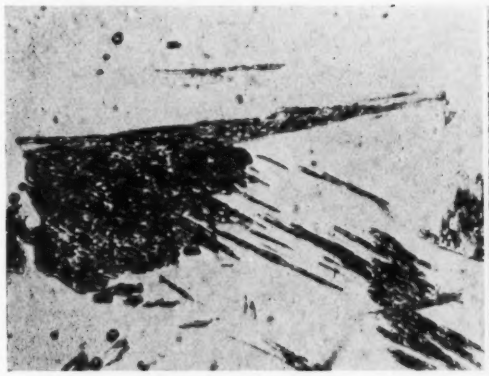

Fig. 5—Bainite in En 19, isothermally transformed at 400° C. for 40 sec. × 800

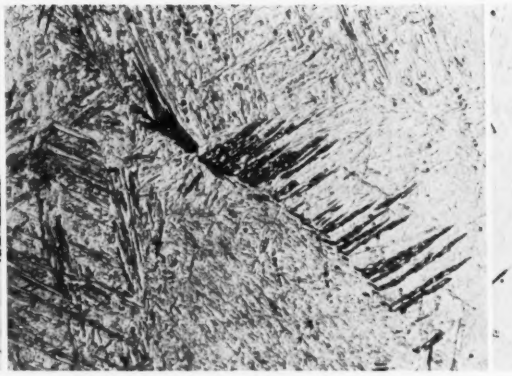

Fig. 6—Bainite in En 17, isothermally transformed at 450° C. for 15 sec., austenitized at 1350° C. × 500

Fig. 7—Bainite in En 25 (VA), isothermally transformed at 510° C. for 80 sec., austenitized at 1250° C. × 150

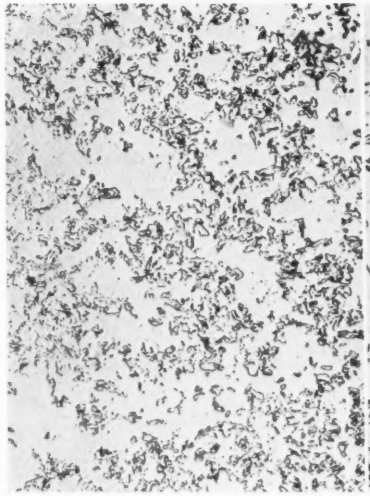

Fig. 8—Bainite in En 25 (VB), isothermally transformed at 465° C. for 200 sec., austenitized at 950° C. × 150

Fig. 9—Bainite in En 25 (VB), isothermally transformed at 465° C. for 100 sec., austenitized at 1250° C. × 150

Fig. 10—Bainite in En 25 (VB), isothermally transformed at 465° C. for 200 sec., austenitized at 1250° C. × 150

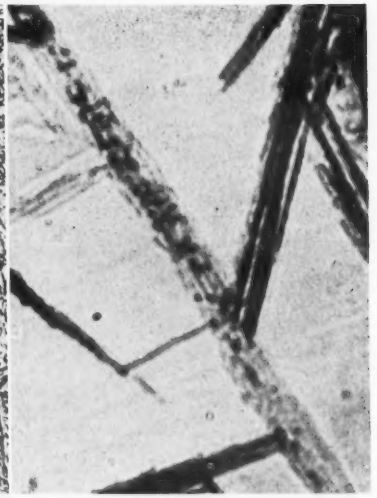

Fig. 11—Bainite in En 31, isothermally transformed at 275° C. for 2 hr., austenitized at 1250° C. × 3000

Temper-Brittleness:

A Critical Review of the Literature

By B. C. Woodfine, B.Met.

SYNOPSIS

A critical survey has been made of the published work on temper-brittleness up to 1952. Particular attention has been paid to those papers published since 1944, and it is shown that several of the suggestions concerning temper-brittleness which have been advanced are not supported by the experimental results. The various theories which have been proposed to explain temper-brittleness are summarized, and it is concluded that further experimental work is required to clarify a number of outstanding points relating to its occurrence and development. 729

HISTORICAL INTRODUCTION

It has been known for some time that the notched-bar impact properties of many quenched and tempered low-alloy steels are adversely affected if, after tempering, the steels are slowly cooled through, or reheated in, the temperature range 600–300° C., and it is to this phenomenon that the name temper-brittleness has been given. The effect of different rates of cooling on the toughness of a quenched and tempered Ni–Cr steel is shown by the following results[1]:

Method of Cooling from 650° C.	Fracture Energy, ft.lb. (Izod test at R.T.)
Furnace	7
Air	17
Oil	44
Water	55

According to Howe,[2] the practice of water-quenching after tempering, to avoid embrittlement on cooling, was used by certain blacksmiths as early as 1883, and was known as water-annealing. Various Krupp patents[3] taken out during 1900 also refer to the desirability of quenching Ni–Cr steels after tempering. However, it was not until 1914–18, when large quantities of alloy steels were used for armour and ordnance, that temper-brittleness received serious attention. The use of the notched-bar impact test as an acceptance test for steels used in the munitions industry made it necessary to heat-treat these steels to obtain high impact values.

The first published reference to temper-brittleness appears to have been made by Brearley[4] in the discussion to a paper which appeared in 1917, and during the same year the term temper-brittleness was coined by Dickenson.[5] Other papers and discussions were contributed by Philpot,[6] Hatfield,[7] Grenet,[8] and Dickenson,[9] and considerable experimental data were published by Greaves and his co-workers[10–13] beginning in 1919. The method used by all these investigators in studying the problem was to compare the room-temperature impact results obtained after different treatments. The liability of a particular steel to temper-embrittlement was expressed as the ratio of the notched-bar impact results after water-quenching and after furnace-cooling, respectively, from 650° C., and this ratio was referred to as the 'susceptibility ratio of the steel.'

In spite of many investigations and a large quantity of experimental data, there was no agreement on the mechanism or cause of temper-brittleness; in fact, wide divergencies were reported by different workers from similar investigations. However, the problem became less important from the practical aspect when in 1925 Greaves and Jones[13] found that the embrittlement could be reduced or removed by the addition of Mo. Although further experimental work was carried out between 1925 and 1944, temper-brittleness remained one of the minor mysteries of metallurgy. By 1939, it was regarded as a form of precipitation-hardening[1].

In 1944, as the result of a paper by Jolivet and Vidal,[14] it became clear that the comparison of the results of room-temperature impact tests was not satisfactory for studying temper-brittleness. In all ferritic steels, the fracture energy of notched-bar impact specimens varies with the testing temperature as in Fig. 1, and Jolivet and Vidal pointed out that

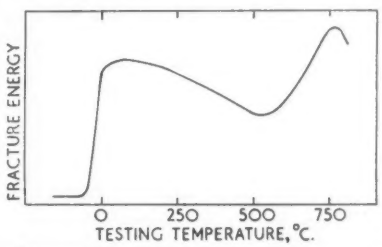

Fig. 1—Schematic variation of fracture energy in notched-bar impact test with testing temperature. Actual shape of curve depends on type of steel and test bar

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XUM

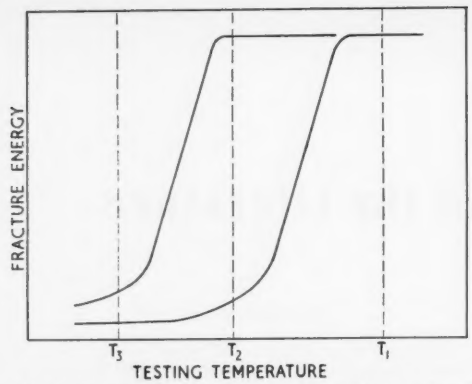


Fig. 2—Low-temperature portion of Fig. 1, showing how temper-brittleness displaces this section of the curve to higher testing temperatures

temper-brittleness did not make the steel uniformly brittle at all testing temperatures, but displaced the low-temperature portion of the curve to higher testing temperatures, as in Fig. 2. It is clear from this that the comparison of notched-bar impact results for only one testing temperature can easily lead to wrong conclusions. In Fig. 2, if the testing temperature is T_1, the steel is apparently not susceptible to temper-brittleness; if it is T_2, the steel is apparently very susceptible, whilst if it is T_3, the steel is apparently only slightly susceptible. Thus, a wide range of values for the susceptibility ratio can be obtained for the same two fracture-energy/testing-temperature curves, depending on their position with respect to the temperature at which the specimens are tested.

The dependence of the notched-bar impact results upon the testing temperature had in fact been known for some time before 1944, and both Monypenny[15] and Greaves and Jones[16] actually published experimental data which showed that temper-brittleness displaced the fracture-energy curve to higher testing temperatures. However, the significance of these results in relation to temper-brittleness was not realized until the publication of Jolivet and Vidal's paper.

It was then necessary to reconsider all the previous conclusions relating to temper-brittleness which had been based on room-temperature impact tests. A full summary and discussion of the work before 1944 has been given by Hollomon,[17] who has attempted to interpret the results and conclusions in terms of the fracture-energy/testing-temperature curves.

In the present review, little attention will be paid to investigations that are confined to the results of room-temperature impact tests, except where they represent the only available information.

EFFECT OF TEMPER-BRITTLENESS ON MECHANICAL AND PHYSICAL PROPERTIES

In investigating the influence of temper-brittleness on the properties of a susceptible steel, no other changes must interfere and produce a spurious effect. The specimens with and without temper-brittleness must be tempered to the same extent. This can be achieved, for example, by quenching two specimens *A* and *B* to produce martensite and then tempering them both at 650° C., after which specimen *A* is furnace-cooled and specimen *B* is water-quenched; both specimens are then reheated to 650° C. and specimen *A* is water-quenched while specimen *B* is furnace-cooled. The properties of the temper-embrittled specimen *B* can then be compared with those of the non-temper-embrittled specimen *A*, in the knowledge that both have received the same amount of tempering.

In previous investigations where such precautions have been taken, although slow cooling from tempering temperatures above 600° C. may have produced a considerable effect upon the notched-bar impact results, it had little or no effect upon the other mechanical properties. Temper-brittleness does not apparently affect the hardness, the yield stress, maximum stress, or elongation, as measured in the tensile test, nor the fatigue properties.[10–13, 17, 18] In severely temper-embrittled steels, the ductility as measured by the reduction in area in the tensile test may be slightly reduced, and some specimens show a longitudinal fracture or a 'star' fracture instead of the more usual cup and cone.[17, 19] The effect upon the notched-bar impact properties is shown in Fig. 2, and the same effect is observed if the notched bars are broken by slow bending.[14, 20]

In regard to the physical properties of susceptible steels, no consistent evidence of any precipitation or transformation in tempered martensitic steels in the range 600–300° C. has been found by thermal analysis, although Grenet[21] reports that Chevenard, using a differential dilatometer, found differences in the thermal expansion between a specimen treated at 500° C. and one quenched from 650° C. to remove temper-brittleness. A similar result was reported by Bischof,[22] who compared specimens water-quenched and furnace-cooled, respectively, from 650° C. However, these results may be due to the relief of internal stresses in the quenched specimens.

Greaves and Jones[12] reported that, allowing for quenching strains, the difference in specific gravity of specimens with and without temper-brittleness was less than 1 part in 20,000. Andrew and Dickie,[23] on the other hand, found that the specific volume depended on the rate of cooling, but it is apparent that most of their steels were tempered above Ac_1.

Grenet[21] stated that there was no difference in the magnetic and electrical properties of temper-brittle and non-temper-brittle specimens, and Greaves and Jones[12] also found that the electrical resistivity was not affected. However, Riedrich,[24] comparing specimens water-quenched and furnace-cooled, respectively, from 650° C., found that the furnace-cooled specimens had a slightly higher electrical resistivity. Although the differences are consistent, there is a considerable scatter in the results and quantitative conclusions are not possible.

Weill,[25] in an X-ray investigation, using a back-reflection technique, reported a distinct difference between the tough and brittle states of a Ni–Cr steel. In the tough condition the (211) K_α doublet was resolved, whilst in the brittle condition it was not. This result is not surprising since the tough state was obtained by tempering a martensitic specimen at 650° C. for 15 min. and water-quenching, and the

brittle state by tempering a similar martensitic specimen at 525° C. for 24 hr. and slow cooling. Thus, the specimens possessed different microstructures and would show differences in carbide particle size, ferrite grain size, and probably in the intensity of the internal stresses. These differences alone would be expected to produce the observed variation in the resolution of the K_a doublet, irrespective of temper-brittleness. This was shown to be the correct explanation by Jacquet, Buckle, and Weill,[26] using a similar back-reflection technique on specimens of one of the Cr steels investigated by Jolivet and Vidal.[14] The specimens had been quenched to produce martensite, tempered at 650° C. for 1 hr., and water-quenched; one specimen was reheated for 24 hr. at 525° C. and again water-quenched. The resolution of the (211) K_a doublet from the temper-brittle specimen was slightly better than from the other specimen—the opposite result to that obtained by Weill for the Ni–Cr steel. The slightly better resolution of the K_a doublet observed by Jacquet, Buckle, and Weill in the temper-brittle specimen of the Cr steel is most probably due to the fact that internal stresses present after quenching the specimen from 525° C. will be lower than those in the specimen quenched from 650° C.

Maloof,[27] using a precision X-ray diffraction method to determine the lattice parameters of tempered martensitic specimens of three Ni–Cr steels, found that the parameters for furnace-cooled specimens were less than those for the corresponding water-quenched specimens; the maximum difference observed was 0·0006 kX. units. He concluded that the difference was probably due to the precipitation of carbide in the slowly cooled (*i.e.*, temper-brittle) specimen.

It appears, therefore, that the only physical or mechanical test which has revealed temper-brittleness is the notched-bar test, carried out either by impact or by slow bending.

MEASUREMENT OF TEMPER-EMBRITTLEMENT AND TEMPER-BRITTLENESS

Jolivet and Vidal[14] defined the position of any fracture-energy/testing-temperature curve by the temperature corresponding to the point of inflexion of the curve, and then took the difference between the values for the specimens with and without temper-embrittlement as a measure of the amount of embrittlement. These principles have been adopted by most subsequent investigators, although alternative definitions have been used to specify the transition temperature of the curves. This may, in general, be defined either with regard to the fracture energy or to the nature of the fracture itself. The transition temperature has been variously defined as the testing temperature at which:

(i) Brittle fracture first appears in the specimens
(ii) The fracture shows a given amount of brittle fracture, usually 50%
(iii) The fracture energy is reduced to some arbitrary value, *e.g.*, 15 or 25 ft.lb.
(iv) The fracture energy is reduced to half the value required for completely ductile fracture.

The conclusions drawn from any set of results will not be greatly altered whichever definition is adopted, provided that the curves to be compared are of the same shape. If they are not, then the conclusions may well depend on the particular definition used.*

In comparing the development of temper-brittleness in different steels, it is generally assumed that the temper-embrittlement as measured by the displacement of the fracture-energy/testing-temperature curves is directly proportional to the extent to which temper-brittleness has proceeded. However, no reasons have been given to show that such a simple relationship exists between temper-brittleness and the temper-embrittlement produced by it.

METALLOGRAPHIC INVESTIGATIONS

Previous workers have agreed that there appears to be no difference in the microstructure of specimens with and without temper-brittleness when they are etched either by the normal reagents, *i.e.*, nital, picral, or by the special carbide reagents such as Murakami's reagent, electrolytic chromic acid, or boiling sodium picrate.[8, 12, 28, 29] However, Cohen, Hurlich, and Jacobson[28] showed that it was possible to distinguish temper-brittle specimens by using an ethereal solution of picric acid to which zephiran chloride† had been added. In the temper-brittle specimens this reagent produced an apparent grain-boundary attack superimposed upon a general etch. The grain-boundary attack in the temper-embrittled steels could not be produced in the same specimens by either Murakami's reagent or by boiling sodium picrate. They showed that the grain-boundary attack was closely related to the presence of temper-brittleness, and succeeded in differentiating between specimens of a Ni–Cr steel when the corresponding fracture-energy curves were displaced by only 20° C.

McLean and Northcott[29] demonstrated that the zephiran chloride reagent was not unique, and that the grain-boundary attack could be produced by a number of reagents based on picric acid.

It has been assumed in both these investigations that the boundaries revealed by the reagents in the temper-embrittled steels were the original austenite grain boundaries, but no direct evidence is available to confirm this assumption. McLean and Northcott[29] reported that sub-boundaries within the grains were also produced by their etchant, and suggested that these enveloped different orientations in the parent austenite grain. Maloof[27] has given some electron micrographs from specimens of three Ni–Cr steels, with and without temper-brittleness, which had been etched in the zephiran chloride reagent. There is an obvious grain-boundary attack in the temper-brittle specimens, but as the replicas were shadowed with Cr and the direction of shadowing was not given, it is not clear whether the etchant has produced a ridge or a groove. In reply to discussion[30, 31] it was stated that "depending upon the brightness and steadiness of the electron beam it was possible for the grain boundaries to appear as ridges or grooves. Merely for convenience, the micrographs were taken when the boundaries appeared as grooves." It is difficult to see how the variations in the electron beam could

* *See* p. 234 *et seq.*

† Zephiran chloride is a mixture of alkyl-dimethyl-benzyl-ammonium chlorides in which the alkyl groups range from C_8H_{17} to $C_{18}H_{37}$.

have reversed the contours of the replica, and the conclusion must be that the shadowing treatment has effectively obscured the real effect of the etching reagent.

FRACTOGRAPHIC RESULTS

Although it was suggested quite early that the brittle fracture in temper-embrittled steels was intergranular, whilst the normal brittle fracture in tempered specimens was cleavage,[6, 32, 33] this view was not universally accepted.[7, 12, 34, 35]

In 1947, McLean and Northcott[29] reported that the brittle fracture in temper-embrittled specimens of a Ni–Cr steel was largely intergranular, but they did not examine a corresponding brittle fracture in the specimens without temper-embrittlement. They associated the intergranular fracture in the temper-embrittled specimens with the grain boundaries produced in the same specimens by the CTAB reagent.

Nutting and Cosslett[36] have used the electron microscope to study the fractures in steels susceptible to temper-brittleness. The surface of an aluminium pressing of the fracture was electrolytically oxidized, and the detached oxide film could then be examined in the electron beam as a replica. Unfortunately, a replica from a ductile fracture of the steel without temper-embrittlement was compared with a brittle fracture of the temper-brittle steel. However, the electron micrographs from the latter clearly showed an intergranular fracture. The replicas from these intergranular surfaces showed markings which indicated the presence of globular particles, but the size distribution of these agreed with that obtained for the carbide particles in the bulk of the steel, and the markings were not therefore considered to be connected with the temper-embrittlement.

Recently, Entwisle and Smith[37] presented two fractographs to demonstrate that a Ni–Cr steel without temper-brittleness showed cleavage facets when fractured at − 180° C., whilst a temper-embrittled specimen of the same steel showed intergranular surfaces.

Orowan[38] considers that typical temper-embrittlement is present in a steel only if the brittle fracture is intergranular, and therefore that the presence or absence of this type of fracture is the essential criterion in deciding whether temper-brittleness has occurred in a steel.

EFFECT OF TIME AND TEMPERATURE

Temper-brittleness may occur in a susceptible steel in the quenched and tempered condition whenever the steel is heated in, or cooled through, the (approximate) temperature range 600–350° C.[4–14] The temper-embrittlement can be removed by reheating the steel to above 600° C. and then quenching to below 350° C. Normal tempering and temper-brittleness will occur simultaneously if a susceptible steel is quenched to martensite and then tempered in the range 350–600° C., and in such cases it is very difficult to separate the effects of each process.

When specimens of a susceptible steel are cooled from 650° C., temper-brittleness occurs over a wide range of temperature, to an extent that depends on the rate of cooling; the slower the rate of cooling the greater the resulting temper-embrittlement. However, it is not possible to disentangle the separate effects of time and temperature from data obtained by varying the rate of cooling; this can only be done if the temper-embrittlement is produced by isothermal treatments.

The first investigation into the isothermal development of temper-brittleness was carried out by Greaves and Jones.[12] Although their impact tests were made at room temperature, they showed that if specimens free from temper-brittleness were held at the same temperature as severely temper-embrittled specimens, the impact values of each eventually became identical. Greaves and Jones also showed that the time to reach this equilibrium increased as the reheating temperature decreased. Using their results and assuming that, for a given decrease in fracture energy, the amount of temper-brittleness would be the same for each temperature, Hollomon[17] plotted a time/temperature/temper-brittleness diagram, analogous to the time/temperature/transformation diagrams used to represent the progress of isothermal transformations. His curves had a simple **C**-shape and showed that for a given reheating time temper-brittleness was a maximum after treatment between 475° and 520° C.

According to Greaves and Jones,[12] the maximum temperature at which temper-embrittlement occurred was 600° C., but some temper-embrittlement may have taken place above this temperature without being detected in their room-temperature impact tests.

Vidal[39] was the first investigator to follow the development of temper-brittleness with time at various temperatures by determining the displacement of the fracture-energy/testing-temperature curves. His results are given in Fig. 3. The specimens were quenched to martensite, tempered at 650° C. for 1 hr., and water-quenched; they were then retempered at various temperatures between 350° and 575° C. and again water-quenched. In a Cr steel (0·25% C, 1·4% Cr) at 525° C., temper-embrittlement occurred very rapidly at first, but after about 5 hr. at this temperature the subsequent embrittlement was small. Similar results were obtained for lower reheating temperatures; both the initial rate of embrittlement and the

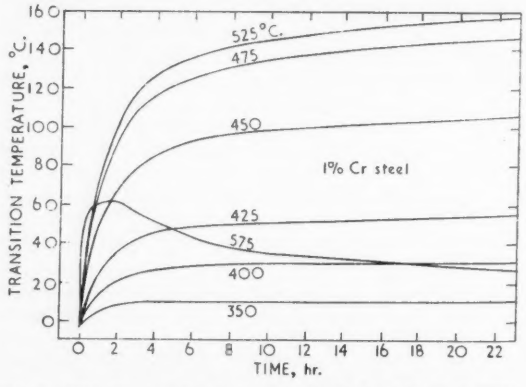


Fig. 3—Vidal's results for a Cr steel (0·25% C, 0·30% Mn, 1·38% Cr, 0·044% P). Transition temperature (determined by reference to an arbitrary fracture energy) plotted against time of treatment at various temperatures (indicated on curves). The steel was originally quenched and tempered at 650° C.

amount of embrittlement decreased with decreasing temperature. At 575° C., however, although the initial rate of embrittlement was greater than that at 525° C., the embrittlement reached a maximum after about 1 hr. and then decreased. Vidal compared this result with the overageing which occurs in age-hardening alloys as the ageing temperature increases.

Jaffe and Buffum[40] studied the temper-embrittlement of a Ni–Cr steel (0·39% C, 0·77% Cr, 1·26% Ni) by a technique similar to that of Vidal, but obtained rather different results (Fig. 4). The steel was austenitized at 900° C. for 1 hr., water-quenched, tempered at 675° C. for 1 hr., and again water-quenched; it was then reheated for various times at various temperatures. The temperature of maximum temper-embrittlement in this steel appears to be about 475° C. for long-time treatments, whilst the rate of embrittlement in the initial stages reaches a maximum, at 550° C. At this latter temperature, the embrittlement apparently decreases after 48 hr.; this overageing was reported in the text but was ignored in drawing the curves. There is no indication of any overageing at temperatures above 550° C., but the amount of embrittlement in this range is very small. Jaffe and Buffum considered that the overageing reported by Vidal was due to softening during the retempering treatments. Unfortunately, no hardness results had been given by Vidal to show that this was not the case.

The time/temperature/temper-embrittlement diagram of Jaffe and Buffum (Fig. 4) is similar to that deduced by Hollomon from the results of Greaves and Jones, except that it shows a second nose at 550° C. However, as already pointed out, the results obtained at this temperature are not consistent with those for the other embrittling temperatures.

The steels used by Vidal and by Jaffe and Buffum were not of the same composition* and this may account for the differences in their results. The amount of temper-embrittlement which results from any given heat-treatment depends very largely upon the composition of the steel, and this probably also applies to the rate of development of the embrittlement and the temperature of maximum embrittlement. This would appear to be borne out by some curves given by Vidal which indicate that a Cr steel, a Mo steel, and a tungsten steel all have different temperatures of maximum embrittlement.

With regard to the maximum temperature at which temper-brittleness can occur, Jaffe and Buffum observed a slight rise in the transition temperature after treatment at 650° C., but also found that appreciable softening had taken place. To reduce this softening in a subsequent experiment, the steel was tempered for 4 hr. at 675° C. before reheating for 48 hr. at 625° and 650° C. The embrittlement at these temperatures was then increased. Furthermore, the transition temperature of the steel after tempering for 4 hr. at 675° C. was higher than after tempering for only 1 hr., and this, in conjunction with the previous results, led Jaffe and Buffum to conclude that temper-brittleness could occur up to the Ae_1 line. However, no micrographs were given to show that temper-brittleness had occurred at these temperatures, and it is possible that the embrittlement, *i.e.*, the increase in the transition temperature, was due to an increase in the ferrite grain size of the tempered martensite. The following later results of Jaffe and Buffum,[41] on the same steel, appear to confirm this suggestion.

Groups of specimens were given the following treatments:

(i) Austenitized 900° C. for 1 hr., and water-quenched, tempered 675° C. for 1 hr., and water-quenched
(ii) As (i), followed by 48 hr. at 500° C. and water-quenched
(iii) As (ii), followed by 1 hr. at 675° C. and water-quenched
(iv) As (ii) + 48 hr. at 675° C. and water-quenched
(v) As (ii) + 240 hr. at 675° C. and water-quenched
(vi) Austenitized 900° C., water-quenched, tempered 675° C. for 240 hr., and water-quenched.

The transition temperatures for these six treatments are shown in Fig. 5; although 1 hr. at 675° C. removed the temper-embrittlement produced by 48 hr. at 500° C., further tempering at 675° C. considerably increased the transition temperature. Jaffe and Buffum compared these results with the reversion phenomenon found in age-hardening, and claimed that they proved that temper-brittleness proceeded by nucleation and growth. However, if Fig. 5 is compared

* The compositions were as follows: Vidal—0·25% C, 0·32% Si, 0·30% Mn, 1·38% Cr, 0·016% S, 0·044% P; Jaffe and Buffum—0·39% C, 0·30% Si, 0·79% Mn, 0·77% Cr, 0·028% S, 0·015% P, 1·26% Ni, 0·02% Mo.

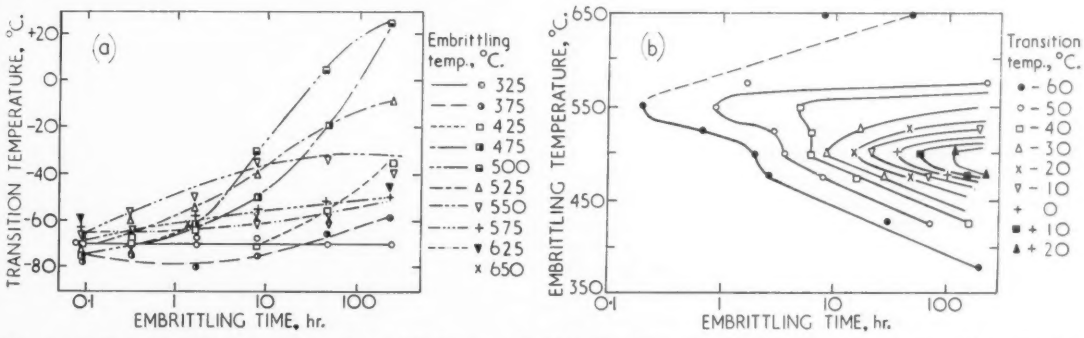
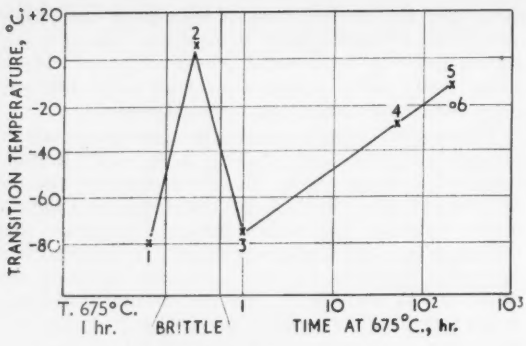


Fig. 4—Jaffe and Buffum's results for a Ni–Cr steel (0·39% C, 0·79% Mn, 0·77% Cr, 0·015% P, 1·26% Ni): (*a*) Transition temperature (first appearance of brittle fracture) plotted against time at various temperatures (log scale); (*b*) time/temperature/embrittlement diagram derived from (*a*)



(1) T. 675° C., 1 hr.
(2) As (1) + 48 hr. at 500° C.
(3) As (2) + 1 hr. at 675° C.
(4) As (2) + 48 hr. at 675° C.
(5) As (2) + 240 hr. at 675° C.
(6) T. 675° C., 240 hr.

Fig. 5—Jaffe and Buffum's results for a Ni–Cr steel (same composition as in Fig. 4) given various treatments after quenching

with Fig. 4, it will be seen that the temper-embrittlement produced by 240 hr. increases from 325° to 500° C., decreases between 500° and 625° C., and then suddenly increases again at 675° C.

These results can only be explained by a mechanism of nucleation and growth if the nature of one of these processes undergoes a radical change between 675° and 625° C.; no mention of this was made by Jaffe and Buffum. There is in fact no metallographic or fractographic evidence that the increase in the transition temperatures at 675° C. is due to temper-brittleness, and it is very significant that for the specimens tempered at this temperature the transition temperatures are closely related to the Rockwell hardness values (*see* Fig. 6). This correlation suggests that the embrittlement at 675° C. is not due to temper-brittleness but probably arises from the increase in the ferrite grain size which occurs as tempering proceeds and the carbide particles coalesce.

Pellini and Queneau[42] found that embrittlement occurred in a quenched and tempered Ni–Cr–Mn steel after reheating up to 650° C., but unfortunately the impact tests were made at only one temperature in the transition range, and it also appears that the specimens had previously been tempered above the Ae_1 line. Cohen and others[28] showed that, in one steel reheated to 630° C., the zephiran chloride reagent produced a grain-boundary attack; this suggests that temper-brittleness had occurred at 630° C.

The lowest temperature at which temper-brittleness takes place appears to be about 350° C. (Vidal,[39] Jaffe and Buffum[40]), or possibly below (Jones)[43] if the times of treatment were prolonged.

Only one comparison has been made between embrittlement resulting from slow cooling and that occurring in isothermal treatments. Jaffe and Buffum[39] cooled a Ni–Cr steel from 675° C. at 17° C./hr., and although the specimens took only 20 hr. to cool from 675° to 325° C. and were in the temperature range corresponding to the nose of the time/temperature/embrittlement curve for only 2 hr., the temper-embrittlement was comparable with that in specimens held for 48 hr. at the temperature of the nose.

From the foregoing it is obvious that, although there is agreement on the general features of the development of temper-brittleness with time and temperature, many details require further consideration. The most important appear to be the maximum temperature at which temper-brittleness can occur, and the possibility of reducing temper-embrittlement by over-ageing with long-time treatments.

EFFECT OF COMPOSITION

Variations in the composition of a susceptible steel may affect either the amount of temper-embrittlement or the kinetics of the embrittlement. Most of the investigations have so far been concerned with the amount of temper-embrittlement which results from a particular treatment and have not attempted to separate the other possible effects.

The situation is further complicated by the fact that the addition of an alloying element may produce an effect either by itself or by virtue of its association with the other elements present. Thus, an element which does not produce temper-brittleness when added to a non-susceptible steel may increase the temper-brittleness present in a susceptible steel. An alloying element may have a secondary effect on temper-brittleness or on the temper-embrittlement, by altering some variable such as the grain size. Hence, interpretation of data on the effects of alloying elements is not always straightforward.

The addition of an alloying element may bring about other changes apart from temper-brittleness when a steel is reheated below 600° C. after quenching from above this temperature. The fracture-energy/testing-temperature curves will then not merely be displaced, as in Fig. 2, but will show other differences.

Plain Carbon Steels: Effect of the Carbon Content of Susceptible Steels

Most previous investigators consider that plain carbon steels with less than 0·5% of Mn are not susceptible to temper-brittleness,[13, 17] but recently Jaffe and Buffum[44] and Libsch, Powers, and Bhat[45] have suggested that they are very susceptible and that even specimens which have been quenched from tempering temperatures above 600° C. have been embrittled during the quench.

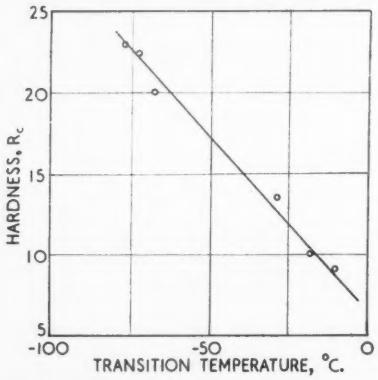


Fig. 6—Results of Fig. 5 with Rockwell hardness plotted against transition temperature (from graphs in the original paper)

Jaffe and Buffum claimed experimental verification for their proposal by comparing the impact properties of a plain carbon steel with those of a Ni–Cr steel. Both steels were quenched and tempered to the same hardness and had the same prior austenite grain size. It was assumed that in the absence of any temper-embrittlement the transition temperatures of the two fracture-energy/testing-temperature curves would be the same. (This is equivalent to assuming that the solution of alloying elements in the ferrite will not affect the impact properties.) They state that "The hypothesis that plain carbon steels develop temper-brittleness so rapidly that they become embrittled even when quenched from tempering temperature, then leads to the predictions that: (i) without the deliberate embrittling treatment the transition temperature of the carbon steel is higher than that of the alloy steel and (ii) with the embrittling treatment the transition temperatures of the two steels are roughly equal." Their results are shown in Fig. 7. The transition temperature was taken as the testing temperature at which the fracture was 50% brittle, and the results were given as follows:

Without Embrittling Treatment:
Plain carbon steel — 85° C.
Alloy steel —115° C. } Difference 30° C.

*After Embrittling Treatment:**
Plain carbon steel — 60° C.
Alloy steel — 55° C. } Difference only 5° C.

* The embrittling treatments were 50 hr. at 455° C. for the carbon steel, and 48 hr. at 480° C. for the alloy steel.

These results agree with the foregoing predictions (i) and (ii), and Jaffe and Buffum thereby concluded that their hypothesis was correct. However, on close examination, the results are not satisfactory. The 'plain carbon steel' contained 0·8% of Mn, which probably gave rise to the slight temper-embrittlement observed. The assumption that the solution of alloying elements in the ferrite does not significantly affect the transition temperature is incorrect, as Hodge, Manning, and Reichhold[46] have shown that, for a given ferrite grain size, Ni added to a 0·02% carbon steel lowered the transition temperature. That the presence of the alloying elements has an effect upon the impact properties is shown by the difference in the shape of the fracture-energy curves for the two steels (*see* Fig. 7). However, considering the curves as given, if the transition temperature is defined as the lowest testing temperature at which the fracture is 100% ductile (as used by Jaffe and Buffum in their other investigations), the following results are obtained:

Without Embrittlement:
Plain carbon steel — 70° C.
Alloy steel — 75° C. } Difference only 5° C.

After Embrittlement:
Plain carbon steel — 60° C.
Alloy steel — 30° C. } Difference 30° C.

The relationships between the transition temperatures are now reversed and consequently do not agree with Jaffe and Buffum's two predictions. The definition of the transition temperature by reference to a 50% brittle fracture has no fundamental significance and

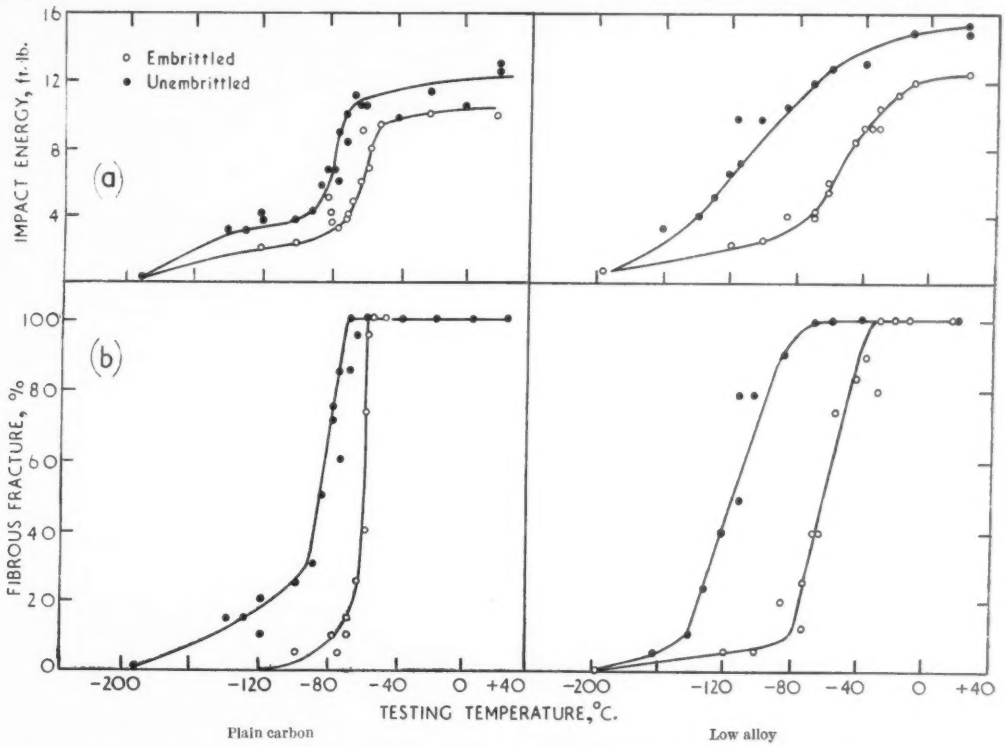


Fig. 7—Jaffe and Buffum's results for a low-alloy steel and a plain carbon steel: Testing temperature plotted against (*a*) fracture energy, and (*b*) percentage fibrous (ductile) fracture in the specimens

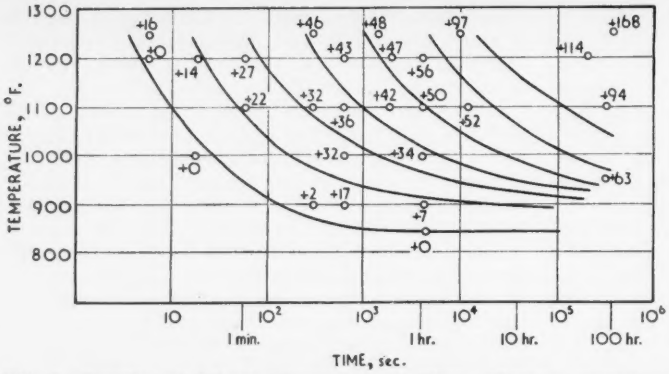


Fig. 8—Results of Libsch, Powers, and Bhat. Figures represent increase in transition temperature (° F.), as compared with unembrittled specimen. Full lines are presumably lines of equal embrittlement

thus their experimental results do not confirm the hypothesis that plain carbon steels develop temper-brittleness very rapidly and are embrittled even after quenching from above 600° C.

Further evidence against this theory is provided by the nature of the brittle fracture obtained in plain carbon steels. Entwisle and Smith[37] found that a 0·4% C, 0·65% Mn steel treated at 480° C. for 24 hr. showed only 5% intergranular fracture when broken at — 180° C., and they concluded that the steel was not susceptible to temper-brittleness in the normal sense of the term. Jaffe and Buffum reported[47] that the fracture in the specimen of the carbon steel without temper-brittleness was chiefly transgranular, but that the fracture appeared to be intergranular in the temper-embrittled specimen. Thus, temper-embrittlement would not appear to be present in the specimen rapidly cooled from above 650° C., and the subsequent development of temper-brittleness in the 'plain carbon steel' can be explained by the high Mn content.

Libsch, Powers, and Bhat[45] decided, from results obtained by Powers, that temper-embrittlement in a plain carbon steel* could only be avoided if the tempering treatment was restricted to very short times and was followed by a rapid quench. They then investigated the isothermal development of temper-embrittlement in this steel. The specimens of the steel without temper-brittleness were obtained by tempering the martensitic specimens at 650° C. for 5 sec. and then water-quenching immediately; this rapid tempering treatment was carried out in an induction heating apparatus. The specimens were then treated for various times at various temperatures. Although considerable softening occurred in these treatments, Libsch, Powers, and Bhat stated that "for the purposes of this investigation the influence of hardness upon transition temperature was not considered significant."

The transition temperatures (based on a 50% fibrous fracture) which were obtained for the various treatments were used to construct an 'iso-embrittlement diagram' for the steel (Fig. 8). This shows that the embrittlement increases with increasing temperature up to the Ae_1 line, and is very small in the range 475–500° C. (887–932° F.). A comparison of Figs. 4 and 8 shows that these results are the opposite of those obtained for low-alloy steels. No information was given on the nature of the brittle fracture in the embrittled specimens, but the specimens did not show a grain-boundary attack when etched in the zephiran chloride reagent of Cohen, Hurlich, and Jacobson. Libsch and his co-workers concluded that temper-brittleness did occur in plain carbon steels, and that it could only be prevented by restricting the tempering time at temperatures above 480° C. to a few seconds. They concluded that the alloying elements reduced the rate of temper-embrittlement above 540° C.

However, if the embrittlement observed in this steel was due to temper-brittleness, it is clear from Figs. 4 and 8 that the alloying elements must not only considerably reduce the rate of temper-embrittlement just below the Ae_1 line, but must also considerably increase it in the range 450–500° C. It is very difficult to see by what mechanism this can be accomplished, and Libsch, Powers, and Bhat did not suggest one. In view of the incompatibility between the results from this plain carbon steel and those from the alloy steel, and the negative response of the embrittled steel to the zephiran chloride reagent, it is reasonable to conclude that the embrittlement observed in the plain carbon steel was not due to temper-brittleness. This is confirmed by Fig. 9, which shows that for this steel there is a definite relationship between the transition temperatures and the Rockwell hardness numbers, irrespective of the tempering temperature. The assumption by Libsch and his co-workers that the hardness did not influence the transition temperatures is thus disproved. The embrittlement at temperatures above 480° C. in the plain carbon steel would appear therefore to be due to the increase in the ferrite grain size which occurs as tempering proceeds.

Libsch and his co-workers agreed[48] that the embrittlement observed in their experiments was predominantly due to increasing ferrite grain size and agglomeration of the carbides in the ferrite grain

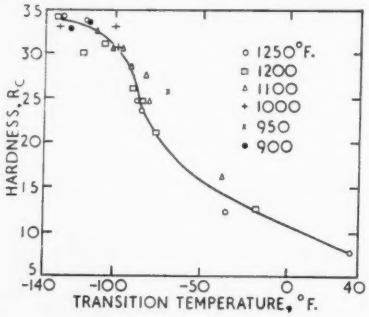


Fig. 9—Results of Fig. 8, with Rockwell hardness plotted against transition temperature for various tempering (or embrittling) temperatures

* The composition was as follows: 0·46% C, 0·75% Mn, 0·034% S, 0·02% P, 0·03% Ni, 0·12% Cr.

boundaries. They also agreed that the form of the 'iso-embrittlement' curves was probably due principally to the growth of the ferrite grains.

Thus, on the available evidence, it appears that the suggestion that plain carbon steels develop temper-brittleness very rapidly is not correct, and that, provided that the Mn is below about 0·5%, they are not susceptible to temper-embrittlement.

No systematic investigation has been carried out to determine the effect of varying the carbon content of a susceptible steel, but Jolivet and Vidal[14] showed that when the carbon content of a Cr steel was reduced from 0·22 to 0·073%, the temper-embrittlement was reduced but not removed. Jaffe[49] is reported to have found that a vacuum-melted steel with 0·016% of carbon was still susceptible to temper-embrittlement, but Buffum, Jaffe, and Clancy[50] subsequently tested an alloy containing 1·5% of Ni, 0·6% of Cr, and only 0·003% of carbon, and showed that it was not susceptible to temper-embrittlement.* The nitrogen was given as 0·0004% and the phosphorus as 0·003%; the alloy did not apparently quench-age but it did show strain-ageing.

Nickel, Chromium, and Manganese

These three elements are considered together, since steels containing two or more of these elements form the largest group of susceptible steels.

Plain Ni steels are not normally considered to be susceptible,[13] but there is little reliable information on this point. McLean and Northcott[29] reported that at the temperature of liquid air the fracture energy of specimens water-quenched from 635° C. was the same as that for those subsequently reheated for 1 hr. at 500° C. No grain-boundary attack comparable with that produced in steels known to be temper-brittle was obtained with their reagent in this Ni steel.

Jones[43] found that, both in a 3% Ni steel and in a 3½% Ni steel, the notched-bar fracture energy (measured at room temperature) decreased considerably when the quenched and tempered steels were reheated at 400° and 450° C. for up to 3000 hr. However, these steels contained respectively 0·15 and 0·24% of Cr, whereas the steel used by McLean and Northcott had only 0·01% of Cr. It is not possible, therefore, to decide from the available results whether plain Ni steels are susceptible to temper-brittleness or not.

Jolivet and Vidal[14] have shown that plain Cr steels are susceptible to temper-brittleness and that the temper-embrittlement produced by a given treatment increases with increasing Cr. Plain Mn steels have also been shown to be susceptible, the temper-embrittlement likewise increasing with the Mn.[51, 52]

In general, in a steel containing Ni, Cr, and Mn, increasing any one of these elements increases the susceptibility.[17, 51] Although it is doubtful whether Ni steels are themselves susceptible, the addition of Ni to a susceptible steel certainly increases the embrittlement in a given treatment.[51] The effect of increasing the Cr content of a steel containing Mn, Ni, and Cr appears to be independent of the other two elements, but Taber, Thorlin, and Wallace[51] consider that the effect produced by Mn or Ni may be increased by the presence of Cr. Hollomon[17] shows that if two steels have the same hardenability with respect to Ni, Cr, and Mn, they will have approximately the same susceptibility.

Molybdenum

This element is exceptional, for, whilst Vidal[39] has shown that a plain 2% Mo steel† was susceptible, Jolivet and Vidal[14] had previously shown that the addition of 0·26% of Mo to a susceptible Cr steel prevented any embrittlement from developing in 24 hr. at 525° C. There is much evidence that the addition of Mo to Cr, Ni–Cr, and Ni–Mn steels considerably reduces the embrittlement resulting from a given treatment.[13, 51, 53–55] Maurer, Wilms, and Kiessler[56] have reported that the addition of Mo in excess of 0·4% increased the susceptibility, but their impact tests were unfortunately carried out only at room temperature, and the conclusions depend on the comparison of susceptibility ratios. Lea and Arnold[19] consider that the addition of Mo decreased the rate of embrittlement rather than the total amount of embrittlement which was possible, but again these conclusions were based on the comparison of susceptibility ratios.

Vanadium and Tungsten

The only available data on vanadium and tungsten were provided by Jolivet and Vidal,[14] who found that the addition of 0·23% of vanadium to a Cr steel increased the embrittlement which occurred in 24 hr. at 525° C., and by Vidal[39], who has shown that a 3·8% tungsten steel with no Cr and low Mn was susceptible to temper-brittleness.

Oxygen, Sulphur, and Silicon

There are no reliable results on the effect of oxygen and silicon, but it has been reported[14] that increasing the sulphur content from 0·016 to 0·037% in a Cr steel has no effect on the temper-embrittlement.

Phosphorus

There has been much controversy over the effect of phosphorus, but early work based on susceptibility ratios cannot be interpreted because of the effect produced by phosphorus on the transition temperature of the steels without temper-embrittlement. Jolivet and Vidal[14] found that increasing the phosphorus content of a Cr steel from 0·012 to 0·044% not only increased the transition temperature of the steel without temper-brittleness but also considerably increased the temper-embrittlement. Baeyertz and co-workers[54, 55] showed that the temper-embrittlement in a 0·4% C, 1·8% Mn steel was progressively increased for a given treatment as the phosphorus was raised from 0·012 to 0·036%, and that the temper-embrittlement in a Cr steel also increased with the phosphorus content. Similar results were obtained by Herres and Elsea[57] for a Mn and a Mn–Cr steel.

Temper-brittleness has, however, been observed in Cr steels with phosphorus as low as 0·008%.[14]

* The alloy was completely ferritic and was water-quenched from 1150° C., then water-quenched from 900° C., tempered at 360° C. for ½ hr., air-cooled, and finally tempered at 675° C. for 1 hr. and furnace-cooled.

† The composition was as follows: 0·11% C, 0·29% Si, 0·33% Mn, 2·1% Mo, 0·010% S, 0·020% P.

Nitrogen

By comparing susceptibility ratios, Griffiths[58] found that embrittlement was produced when nitrogen was added to plain carbon steels, but it is not certain whether this was due to temper-brittleness. According to Hollomon,[59] temper-brittleness occurred in some experimental alloys with 0·002% of nitrogen. Recently, Imai and Ishizaki[60] claimed that temper-brittleness was due to the precipitation of iron nitride from solution in the ferrite, but as their impact tests were made only at room temperature and the effects of normal tempering and temper-brittleness (if present) were inextricably confused, the true effect of the nitrogen additions is not clear.

Aluminium, Boron, Titanium, and Zirconium

An addition of 0·5% of Al to a Cr steel did not affect the susceptibility.[14]

Kishkin[61] reported that the susceptibility, as measured in room-temperature tests, was decreased by the addition of Ti, but the effect was less marked as the austenitizing temperature was increased.

According to Herres and Lorig,[62] boron steels are susceptible, and Ti and Zr do not prevent temper-embrittlement, but no evidence was produced to substantiate these statements.

Antimony

Jolivet and Vidal[14] reported that a Cr steel containing 0·08% of Sb was severely embrittled by reheating at 525° C. for 24 hr., and that after this treatment brittle fracture was observed at temperatures up to 525° C. The temper-embrittlement could be followed by tests at 525° C. No further data are available on the effect of Sb. It is most probable that it produces a specific embrittlement that is not connected with temper-brittleness.

EFFECT OF MICROSTRUCTURE

Most of the work on temper-brittleness has been concerned with the temper-embrittlement of steels heat-treated to produce tempered martensitic microstructures, and there is little information on the development of temper-brittleness in other microstructures. The first reference to temper-brittleness in unhardened steels was made by Dickenson,[9] and both Grenet[8] and Greaves and Jones[12] obtained results which indicated that temper-embrittlement could be produced in unhardened steels but that the susceptibility ratios were less than those for the quenched and tempered steels. Pellini and Queneau[42] gave some incomplete fracture-energy curves for specimens temper-embrittled after various isothermal treatments. However, only one group had been completely transformed and the temper-embrittlement in these pearlitic specimens was less than in the tempered martensitic specimens after the same treatment, although the complete results could not be obtained from the curves.

According to Greaves and Jones,[12] and to Maurer, Wilms, and Kiessler,[56] increasing the austenite grain size increased the susceptibility in room-temperature impact tests. Recently, Hurlich[63] has given some fracture-energy curves which indicate that the temper-embrittlement was greater when the austenite grain size was increased.

THEORIES

Many theories have been advanced to explain temper-brittleness, which has been variously attributed to:

(i) A transformation below 700° C., Le Chatelier[64]
(ii) An allotropic modification of iron, Jeffries[65]
(iii) Precipitation of a compound from solution in the ferrite; the suggested compounds include: (*a*) cementite or another carbide, Rogers,[66, 67] Grenet,[68] Andrew and Dickie,[23] Honda and Yamada,[69] Nagasawa,[70] Jaffe and Buffum,[41] Kishkin[61]; (*b*) chromium oxide, Greaves and Jones[12]; (*c*) various nitrides, Griffiths,[58] Hollomon,[17] Imai and Ishizaki[60]; (*d*) various phosphides, Feszenko and Czopiwski,[71] Bennek.[72]
(iv) Decomposition of retained austenite, Esser and Eilander,[73] Spretnak and Speiser[74]
(v) Modification of special carbides, Maurer and Hohage,[75] Bischof,[22] Maurer, Wilms, and Kiessler[56]
(vi) Segregation of various elements to the grain boundaries, McLean and Northcott[29]
(vii) Segregation of particles of another phase to the ferrite grain boundaries, Jaffe[76]
(viii) Distribution of the carbides, Maloof.[27]

The first two of these suggestions are now only of historical interest, and of the remaining theories the precipitation hypothesis has been most favoured. However, no precipitate has been detected in the temper-embrittled steels and the micrographs showing the effect of the zephiran chloride reagent on the temper-embrittled specimens do not reveal a distinct grain-boundary phase but rather a grain-boundary attack. To overcome this, McLean and Northcott[29] suggested that the embrittlement might arise from a segregation of various elements, particularly carbon, to the grain boundaries; however, this suggestion has not met with much agreement.

Although Jaffe[76] has demonstrated that for certain values of the various interfacial energies small particles of a phase may segregate to the grain boundaries, there is no experimental evidence to support this theory.

Recently, Spretnak and Speiser[74] have revived the retained-austenite theory and have suggested that, as the result of carbon segregation, austenite was retained at the grain boundaries on quenching from above Ae_3 and subsequently decomposed to give films of martensite at the grain boundaries. However, this ingenious suggestion does not explain how temper-brittleness can be produced in susceptible steels by slow cooling from 650° C. after tempering sufficiently to decompose any retained austenite. According to the theory of Spretnak and Speiser, the retained austenite is required to behave in a manner which is completely incompatible with that which would be expected from the results of Cohen and his co-workers.[77]

Recent work on the composition and stability of alloy carbides indicates that carbide decomposition of the type suggested by Maurer and others is not likely to occur.

CONCLUSIONS

The main conclusion to be drawn from this review is that, although the outlines of the problem of temper-brittleness are clear, many aspects require further investigation and clarification before a satisfactory theory can be agreed upon.

There is a danger that the problem may become obscured by differences in nomenclature. It has been suggested, for example, that any embrittlement (as indicated by an increase in the transition temperature of the fracture-energy/testing-temperature curve) which occurs on tempering should be called temper-embrittlement, irrespective of the cause or nature of the changes involved.[45] This might well lead to confusion, and a better approach would appear to be to determine the nature and characteristics of the embrittlement in, say, Ni–Cr steels (to which the name temper-brittleness was originally given) and then to restrict the term temper-brittleness to those cases where the same features are observed.

The facts which may be regarded as established by previous work are as follows:

(1) The embrittlement occurs in steels containing Cr, Mn, W, Mo, and Ni, although steels which contain only Ni do not appear to be susceptible, and Mo in combination with the other elements may reduce the susceptibility. Sulphur and Al do not appear to affect temper-brittleness, and it does not occur in plain carbon steels with less than 0·6% of Mn.

(2) Temper-brittleness displaces the low-temperature portion of the fracture-energy/testing-temperature curve to higher testing temperatures, but does not appear to affect the other mechanical properties of the steels. It has very little effect, if any, on the physical properties of the susceptible steels.

(3) Temper-brittleness cannot be detected by any of the normal etching reagents, although it is revealed by etchants based on picric acid and containing small traces of surface-active compounds.

(4) Temper-brittleness occurs in steels with a tempered martensitic microstructure, either on slow cooling from 650° C. or on reheating in the range 350–650° C. It reaches a maximum after treatment between 475° and 525° C. Temper-brittleness does not occur to any appreciable extent above 600° C. and can be removed from a susceptible steel by quenching from about 650° C.

The aspects of temper-brittleness which require further investigation include:

(1) The effects of such alloying elements as P, N, Ti, Co, V, and B.

(2) The effect of temper-brittleness on the low-temperature tensile properties of the susceptible steels.

(3) The nature of the etch produced by the reagents which reveal temper-brittleness, and the relationship between the etch and the normal microstructure of the steels.

(4) The possibility of a decrease in the temper-embrittlement with long-time treatments above 500° C., and also the maximum temperature at which temper-brittleness can occur.

(5) The occurrence of temper-brittleness in microstructures other than tempered martensite, and the effect of such variables as grain size on the development of the embrittlement.

(6) The effect of temper-brittleness upon the nature of the ductile and brittle fractures in the susceptible steels.

Until further investigations have settled some of the disputed features of temper-brittleness, a detailed consideration of the various theories proposed to explain the phenomenon would be of little value.

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Some Aspects of Temper-Brittleness

By B. C. Woodfine, B.Met.

SYNOPSIS

Several features of the occurrence and development of temper-brittleness in low-alloy steels have been investigated, including the development of the embrittlement with time and temperature, the occurrence of the embrittlement in isothermally transformed microstructures, the effect of austenitizing treatment, and the effect of the alloying elements aluminium, phosphorus, and titanium. The etching attack produced by various reagents in temper-brittle steels occurs principally along the previously existing austenite grain boundaries, and consists of a network of grooves. No evidence has been obtained of a grain-boundary precipitate.

Brittle fracture in temper-brittle steels has been shown to occur by intergranular failure and not by cleavage. The notched-bar impact test has been examined theoretically and an attempt has been made to relate it to the fundamental flow and fracture properties of the steels. From the proposed relationship it was predicted that temper-brittleness lowered the brittle-fracture strength of a susceptible steel without affecting the yield stress, and this has been confirmed by low-temperature tensile tests. A theory is suggested to explain the occurrence of temper-brittleness.

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Introduction

Previous Work

PRIOR to 1944, temper-brittleness was usually studied by comparing the room-temperature impact values obtained after different treatments, the susceptibility of a given steel to temper-embrittlement being expressed as the ratio of the impact values after water-quenching and after furnace-cooling respectively from 650° C. That this was unsatisfactory was clearly shown in 1944 by Jolivet and Vidal,[1] who demonstrated that the effect of temper-brittleness on the notched-bar impact properties of a susceptible steel was to displace to higher temperatures the transition from ductile to brittle fracture, which occurs in all ferritic steels. As a result of their work it became necessary

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to reconsider all the previous conclusions relating to temper-brittleness, which were based on the results of room-temperature impact tests.

A summary and discussion of previous work on temper-brittleness has been given elsewhere.[2]

Effect of Temper-Brittleness upon Notched-Bar Impact Results

In all ferritic steels the energy required to fracture a notched-bar impact specimen varies with the testing temperature as shown by curve *A*, Fig. 1; but in the steels susceptible to temper-brittleness, if curve *A* represents the results obtained from specimens water-quenched from 650° C., then curve *B* represents those obtained from specimens furnace-cooled from 650° C. or retempered below 600° C. As a result of the slow cooling or the retempering, the whole fracture-energy/testing-temperature curve has been displaced to higher testing temperatures. Impact specimens from the water-quenched steel show a completely ductile fracture at testing temperatures above T_1, a completely brittle fracture at testing temperatures below T_2, and a fracture which includes both ductile and brittle failures at testing temperatures between T_1 and T_2. Corresponding specimens from the temper-brittle steels show fractures that are ductile above T_1', brittle below T_2', and mixed between T_1' and T_2'.

Figure 1 shows that the comparison of impact results obtained for only one testing temperature may give a completely misleading picture of the extent of the temper-embrittlement. For example, if the testing temperature is above T_1' there will be no sign of any embrittlement at all. To obtain a true picture of the effect of temper-brittleness in a given steel, it is necessary to carry out the impact tests over a range from above T_1' to below T_1—the point that was overlooked in the early investigations and that was brought out by the work of Jolivet and Vidal.

Although at suitable testing temperatures brittle fracture can occur in specimens with and without temper-embrittlement, in the water-quenched specimens it takes place by cleavage and in the temper-brittle specimens by intergranular fracture. Recently it has been suggested that temper-brittleness should be defined as any embrittlement (*i.e.*, any displacement of the fracture-energy/testing-temperature curve to higher testing temperatures) which developed progressively with time during the tempering of steel, regardless of the source or nature of the embrittlement.[3] However, in the present work a steel will only be considered to show temper-embrittlement if the displacement of the fracture-energy curves is associated with a partial or complete change in the nature of the brittle fracture from cleavage to intergranular.

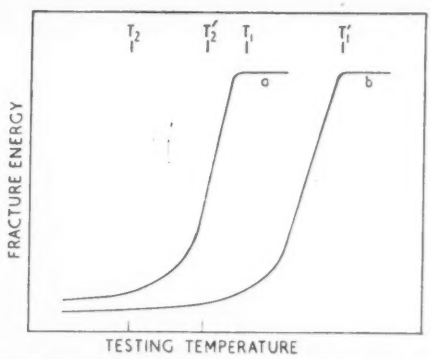


Fig. 1—Effect of temper-brittleness upon the notched-bar impact results of a susceptible steel

After careful consideration of the results of previous investigators, it was decided that, although the notched-bar impact test was open to serious objections regarding the interpretation of the results, it was the most suitable test for studying temper-brittleness in anything like a quantitative manner. Thus the greater part of the present investigation has been carried out using this test.

The theory of the notched-bar test has been examined in detail and an attempt has been made to relate it to the fundamental flow and fracture properties of the steels, and thus to explain the effect of temper-brittleness upon these properties.

EXPERIMENTAL MATERIALS AND METHODS

Alloys

The analyses of the steels used in this investigation are given in Table I. Alloys *J* and *K* were commercial steels of unknown manufacture; steels *B*, *C*, *D*, *G*, and *H* were made by the crucible process in 60-lb. melts, and steels *T* and *U* in a high-frequency induction furnace in 15-lb. melts. All the steels were rolled or forged to ½-in. square bars.

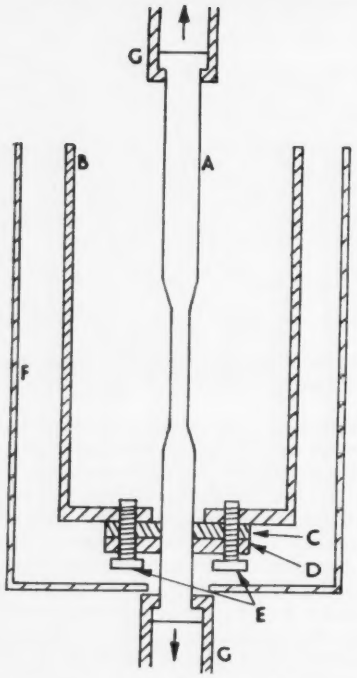
Heat-Treatments

In view of the difficulty in interpreting the results of continuous cooling experiments, temper-brittleness has been produced in the present investigation entirely by isothermal treatments. All the specimens were

Table I

ANALYSES OF STEELS USED

Steel	C, %	Mn, %	P, %	S, %	Si, %	Ni, %	Cr, %	Al, %	Ti, %	N, %
B	0·41	0·38	0·030	0·033	0·19	3·33	1·22	nil	nil	0·0102
C	0·35	0·38	0·029	0·031	0·30	3·29	1·06	0·12	0·47	0·0040
D	0·35	0·38	0·031	0·034	0·32	3·21	1·04	0·29	1·50	0·0051
G	0·36	0·45	0·032	0·032	0·42	0·30	0·12	0·58	3·20	0·0020
H	0·51	0·66	0·031	0·030	0·28	3·24	1·20	0·95	nil	0·0252
J	0·51	0·38	0·014	0·013	0·20	0·06	2·96	...	...	...
K	0·33	0·59	0·037	0·031	0·27	2·92	0·87	...	...	...
T	0·28	0·08	0·010	0·017	0·26	3·00	tr.	...	...	...
U	0·27	0·08	0·134	0·017	0·25	2·94	tr.	...	...	...



A—Specimen
B—Container for liquid nitrogen
C—Leather washer
D—Split brass disc
E—Brass screws
F—Outer container holding lagging
G—Tensometer grips

Fig. 2—Diagram of the specimen and container for low-temperature tensile tests

water-quenched from the tempering and embrittling temperatures, and this fact will not be repeated with each treatment subsequently described.

Measuring Temper-Embrittlement

Jolivet and Vidal[1] defined the position of any fracture-energy/testing-temperature curve by the temperature corresponding to the point of inflexion of the curve, and then took the difference between the values for the specimens with and without temper-brittleness as a measure of the amount of embrittlement that had occurred. The principles of this method have been adopted by most subsequent investigators, although several alternative definitions have been used to specify the transition temperature of the curves.

In the present work the transition temperature is defined as the lowest testing temperature at which the fracture of the test bar is 100% ductile. The theoretical basis of this definition will be apparent from the subsequent section on the theory of the notched-bar impact test, but from the practical point of view it has the advantage that the transition temperature can be fairly accurately determined by visual inspection of the fractured surfaces of the test bars. The transition temperature is indicated on all the fracture-energy curves by a vertical arrow.

The temper-embrittlement is taken as the relative displacement of the transition temperature of the fracture-energy/testing-temperature curves and is given by $\Delta\theta = \theta_1 - \theta_2$, where θ_1 and θ_2 are the two transition temperatures.

Notched-Bar Impact Testing

The notched-bar impact tests have been carried out on standard V-notch Charpy specimens (10 × 10 × 55 mm.; notch depth 2 mm., included angle 45°, root radius 0·01 in.), over the range − 200° to + 200° C., using a 120-ft.lb. Amsler Universal Impact Testing Machine, which had a calculated striking velocity of 16·0 ft./sec. The specimens were heated or cooled to the required testing temperature by immersion in an appropriate liquid, and after soaking for 20 min. were removed and fractured within 3 sec. Tests with a specimen drilled to hold a thermocouple showed that the temperature change during this period was less than 2° C. The temperatures between − 200° and 0° C. were measured with a copper-constantan thermocouple, and those between 0° and + 200° C. with mercury thermometers.

With care in the preparation and testing of the specimens, the number required to plot a fracture-energy/testing-temperature curve and to determine the transition temperature could be reduced to about 10. Thus each point on the various curves corresponds to one specimen only.

Low-Temperature Tensile Testing

The tensile tests in liquid nitrogen were carried out using a vertically suspended Hounsfield tensometer, with the arrangement of the specimen and container shown in Fig. 2. The minimum cross-sectional area of the specimen was approximately 0·02 sq. in.

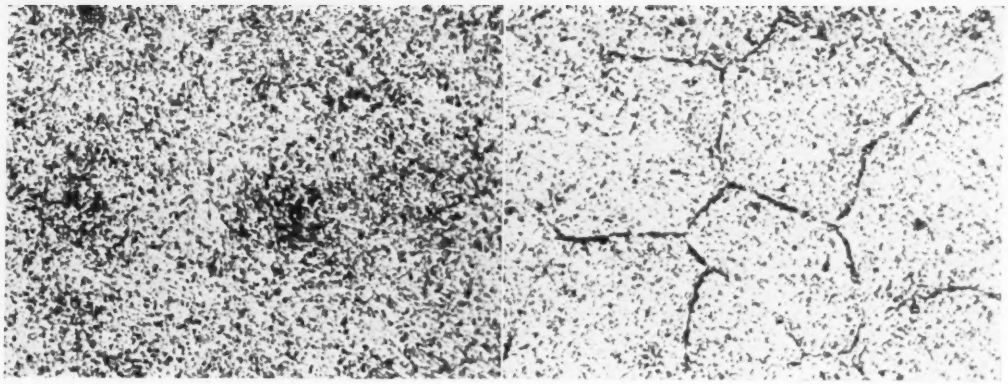
Etching Reagents

In agreement with the results of previous investigators[4–7] no differences in the microstructure of specimens with and without temper-brittleness could be detected in any of the steels examined using the normal etching reagents, but a difference could be produced with an ethereal solution of picric acid containing a surface active compound, in this case cetyl tri-methyl ammonium bromide (CTAB). Subsequently, it was discovered that a saturated solution of picric acid in distilled water would produce exactly the same effect as the CTAB reagent, and this was adopted as the standard etchant.

The specimens to be etched were polished in the normal manner and given a light etch in picral; this etch was polished off and the specimens were then etched for about 10 min. in the aqueous solution. Wherever possible the specimens to be compared were mounted together and polished and etched under the same conditions.

Electron Microscope Technique

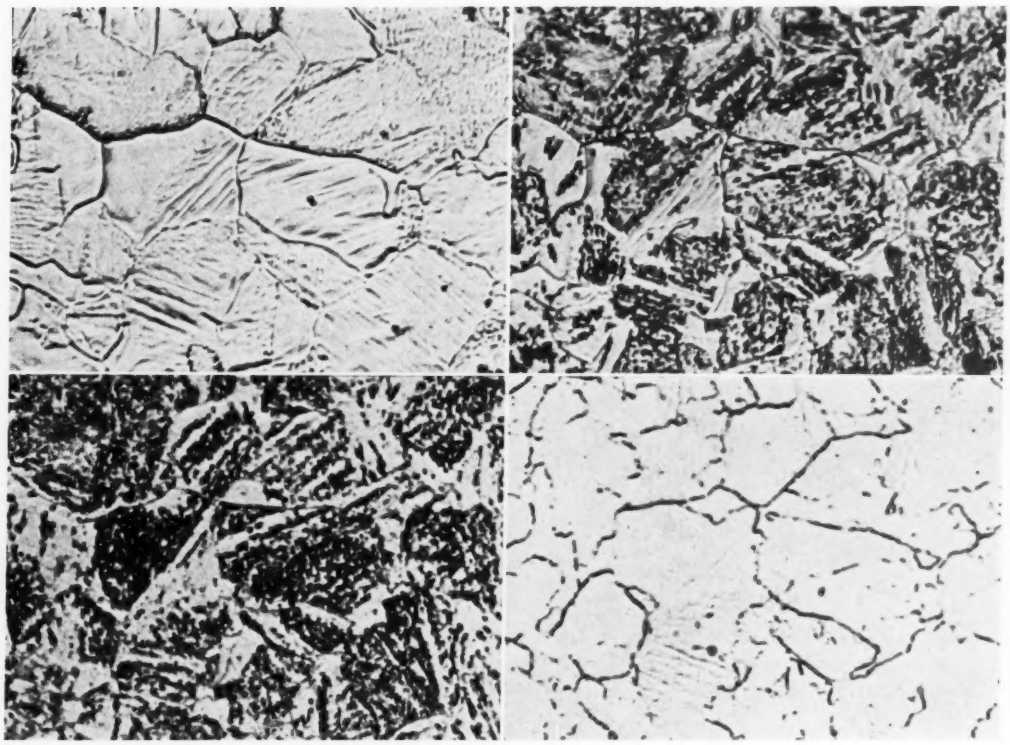
The etched surface of the specimen was first cleaned by the application of a 2% solution of Formvar in chloroform, which was allowed to dry and was then stripped off with adhesive tape. The solutions used to obtain the replicas were ½, 1, and 1½% solutions of Formvar in either chloroform or dioxan. The solution was dropped on to the etched surface and the specimen was then held vertically to allow the liquid to drain and to dry. In some cases the replica was obtained by dry-stripping the film directly on to



(*a*) W.Q. 650° C. (*b*) Furnace-cooled 650° C.

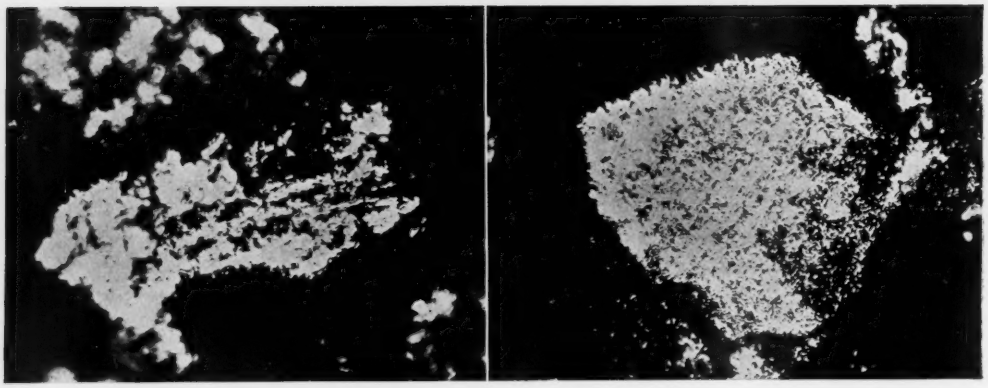
Fig. 3—Steel *J*, tempered martensite etched with aqueous picric acid × 700

(*a*) Thermal etch (*b*) Thermal etch with light picral etch superimposed



(*c*) T. 650° C. and furnace-cooled ; picral etch (*d*) Etched with aqueous picric acid, with background removed by a light polish

Fig. 4—Steel *K*. The same specimen area after various treatments × 1100



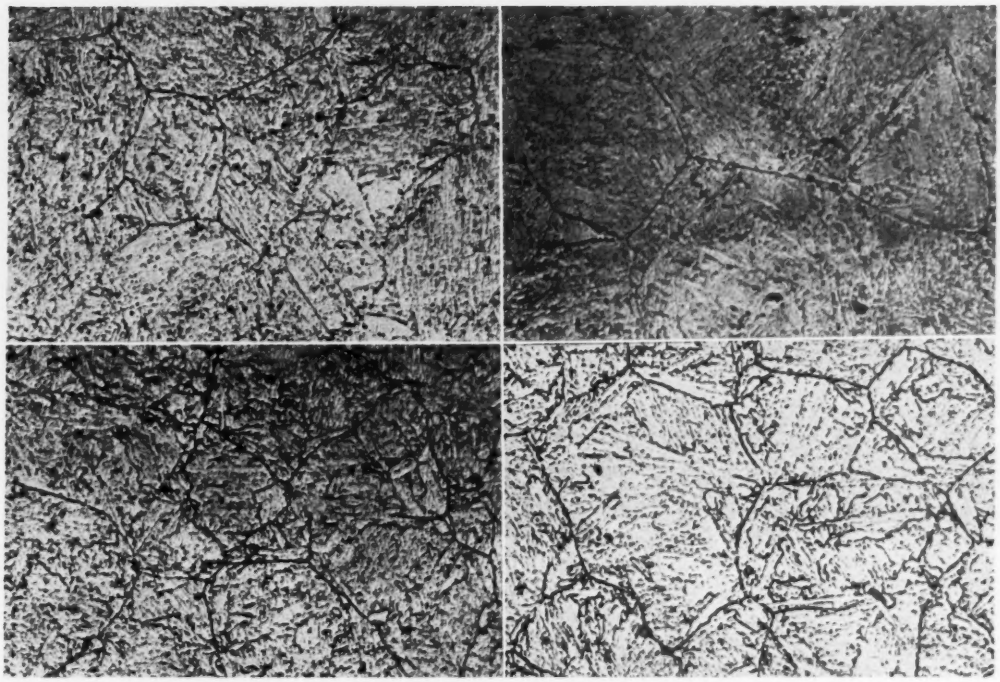
(*a*) Specimen without temper-embrittlement, showing cleavage facets (*b*) Temper-embrittled specimen, showing intergranular facet

Fig. 5—Fractographs of brittle fractures in steel *K* × 400

Fig. 6—Electron micrograph of temper-brittle specimen of steel *K*. Aqueous picric acid etch, 3 min. × 10,000

(*a*) Untreated ; transition temp. −59° C.

(*b*) ¼ hr. at 500° C. ; transition temp. −18° C.



(*c*) ¼ hr. at 500° C. ; transition temp. +17° C.

(*d*) 2 hr. at 500° C. ; transition temp. +68° C.

Fig. 7—Effect of aqueous picric acid; 5-min. etch on specimens of steel *K*, treated for various times at 500° C., after T. 1 hr. at 650° C. × 750

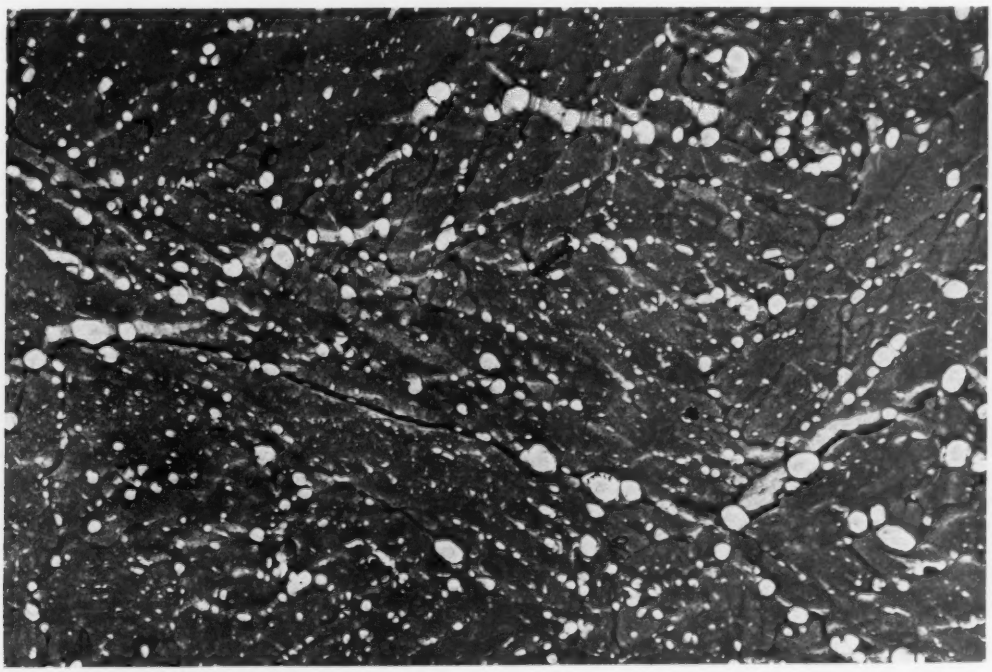


Fig. 8—Electron micrograph of specimen of steel *K*, treated for 128 hr. at 575° C. after T. 1 hr. at 650° C. × 7250

XUM

(b)

(a)

Fig. 9—Electron micrographs from temper-brittle pearlitic specimen of steel *K*, etched with aqueous picric acid: (**a**) for 3 min.; (**b**) for 5 min. × 7500

a grid, but generally the film was backed with Solvar and then dry-stripped. In the latter case the replica was separated from the backing film by soaking in water, and was then transferred to the grids by the normal methods used in wet-stripping.

With negative plastic replicas the interpretation of the micrographs is based on the assumption that the top surface of the film is flat and that the differences in thickness truly represent the surface contours of the specimen. Thus, as the replicas are examined by a transmitted beam of electrons, the ' valleys ' in the original specimen, which have given rise to thick areas in the film, will appear dark in the final micrograph, and the ' hills ', which have produced thin areas in the film, will appear light.

EXPERIMENTAL RESULTS

Metallographic Investigations

The effect of aqueous picric acid on specimens of a susceptible steel water-quenched and furnace-cooled from 650° C. is shown in Fig. 3. In the temper-embrittled specimen the etchant has produced some form of grain-boundary attack superimposed on the general etch present in both specimens. In all the steels examined the grain-boundary attack was either absent or only faintly present in specimens that had been water-quenched from 650° C., but it was well developed in specimens that had been slowly cooled from this temperature. Specimens quenched from 650° C. and subsequently reheated in the temperature range 300–600° C. showed a distinct grain-boundary etch, the extent and intensity of which depended on the reheating time and temperature. When any of the specimens showing the grain-boundary etch were reheated to 650° C. and then water-quenched, the grain-boundary etch was removed.

Thus, from the results outlined above, it appears that there is a close relationship between temper-brittleness and grain-boundary attack.

Relation of Aqueous Picric Acid Etch to Normal Microstructure—The size of the grains revealed by aqueous picric acid is considerably greater than that usually reported for ferrite grains,[8] but whilst it has been assumed by previous investigators that the etched boundaries correspond to the prior austenite grain boundaries,[6, 7] there is no direct evidence to confirm this. Attempts were therefore made to reveal the prior austenite boundaries by a thermal etch *in vacuo* at 900° C. and to compare them with those produced by the aqueous picric acid after the specimen had been subsequently tempered and embrittled.

The thermal etch was produced by heating an electropolished specimen of steel *K* to 900° C. for 1 hr. under a pressure of 10^{-3} mm. Hg, after which the specimen was rapidly cooled in the furnace tube while still keeping it under a reduced pressure. The isothermal transformation characteristics of steel *K* were such that during this cooling the austenite transformed to a bainitic microstructure between 400° and 350° C. Several representative areas of the specimen, located with reference to a number of scribed rings, were then photographed both before and after a light picral etch. The specimen was subsequently heated to 650° C. under reduced pressure for 1 hr. and allowed to cool slowly in the furnace. After all traces of the previous etch had been removed, the specimen was etched in picral and the various areas were rephotographed. The picral etch was removed by a light repolish and the specimen was then etched in aqueous picric acid; after another light repolish to remove the background etch, the specimen was again photographed. The sequence of micrographs obtained for one area is given in Fig. 4, and if allowance is made for the metal removed in polishing, the main grain boundaries produced by the aqueous picric acid correspond to the prior austenite grain boundaries revealed by the thermal etch. This observation was confirmed for the other areas examined. In a number of places an attack has been produced by the aqueous picric acid which does not correspond to any of the austenite grain boundaries, but comparison of the various micrographs reveals that these ' sub-boundaries ' generally correspond to obvious ferrite grain boundaries.

Nature of the Grain-Boundary Etch—At low magnifications, the grain boundaries produced in temper-embrittled specimens by aqueous picric acid had the appearance of films of a dark-etching constituent, but at higher magnifications no such films were visible and the boundaries resembled grooves.

In an attempt to determine the nature of the grain-boundary etch, negative Formvar replicas obtained from lightly etched temper-embrittled specimens were examined in the electron microscope. Figure 6 shows that the etchant has produced a series of grooves in the specimens, and that the etch is not due to the presence of films or particles of a separate phase standing proud from the matrix. Both large and small particles (presumably carbides) lie in the grain boundaries, but in no temper-embrittled specimens were continuous grain-boundary films found. Within the main boundaries, the etchant has revealed a number of sub-boundaries which, in general, do not appear to have been so deeply etched. From their size and distribution relative to the austenite grains, and from the previous section, it is clear that these sub-boundaries are in fact ferrite grain boundaries.

Increasing the etching time made the grain-boundary grooves wider and more continuous.

Fractographic Investigations

Brittle fractures of specimens with and without temper-embrittling treatments were carefully examined, and it was immediately evident that they were not of the same type. The differences are illustrated in Fig. 5. In the steels that were not temper-embrittled the brittle fracture occurred by cleavage across the ferrite grains, each of which gives rise to one of the small bright flat facets shown in Fig. 5*a*. The fact that a number of these facets apparently lie in the same plane indicates that the orientation differences between these grains are small. In the temper-embrittled specimens the brittle fracture generally consisted of large smoothly curved facets which had the characteristic appearance of intergranular surfaces (Fig. 5*b*). Examination of sections at right-angles to the fracture (which had previously been plated with nickel to prevent it from being damaged during polishing and etching) showed that

Table II

DEVELOPMENT OF TEMPER-BRITTLENESS AT VARIOUS TEMPERATURES: STEEL *K*

Treatment*	Transition Temp., °C.	Embrittlement, °C.	V.D.H. (50 kg.)
O.Q., T. 1 hr. 650° C.	− 59		280
O.Q., T., emb. ¼ hr. 500° C.	− 18	41	280
„ „ „ ½ hr. 500° C.	+ 17	76	280
„ „ „ 2 hr. 500° C.	+ 68	127	280
„ „ „ 8 hr. 500° C.	+109	168	279
„ „ „ 32 hr. 500° C.	+137	196	278
„ „ „ 128 hr. 500° C.	+145	204	276
O.Q., T. 1 hr. 650° C.	− 59		276
O.Q., T., emb. ¼ hr. 550° C.	+ 15	74	274
„ „ „ ½ hr. 550° C.	+ 21	80	272
„ „ „ 2 hr. 550° C.	+ 33	92	271
„ „ „ 8 hr. 550° C.	+ 36	95	272
„ „ „ 32 hr. 550° C.	+ 32	91	266
„ „ „ 128 hr. 550° C.	+ 15	74	262
O.Q., T. 1 hr. 650° C.	− 59		282
O.Q., T., emb. ¼ hr. 575° C.	+ 7	66	278
„ „ „ ½ hr. 575° C.	+ 13	72	277
„ „ „ 2 hr. 575° C.	+ 11	70	278
„ „ „ 8 hr. 575° C.	+ 5	64	275
„ „ „ 48 hr. 575° C.	− 12	47	270
„ „ „ 128 hr. 575° C.	− 20	39	259
O.Q., T., emb. ¼ hr. 600° C.	− 24	35	279
„ „ „ ¼ hr. 630° C.	− 50	9	279
O.Q., T., emb. 8 hr. 500° C., then 8 hr. 550° C.	+ 36	95	268
O.Q., T., emb. 8 hr. 550° C., then 8 hr. 500° C.	+109	168	268
O.Q., T., emb. 128 hr. 575° C., then ½ hr. 500° C.	+ 15	74	259
O.Q., T., emb. 128 hr. 575° C., then 2 hr. 500° C.	+ 37	96	258
O.Q., T. 5 hr. 650° C.	− 67		253

* All the tempering and embrittling treatments were followed by water-quenching.

the intergranular facets were associated with the main grain boundaries produced by the aqueous picric acid. This meant that the fracture was following the path of the prior austenite grain boundaries.

The brittle fracture in the temper-embrittled specimens was not always completely intergranular, and specimens that had not been severely embrittled sometimes showed cleavage facets. As the degree of temper-embrittlement increased, the number of cleavage facets decreased. Except in the case of steel *K*, in the coarse-grained condition, intergranular facets were not found in the specimens that were not temper-embrittled.

Effect of Time and Temperature on the Development of Temper-Brittleness

Development of Temper-Brittleness at 500°, 550°, *and* 575° *C.*—Bars of steel *K* (*see* Table I) were oil-quenched from 850° C. and tempered at 650° C. for 1 hr. Various groups of these bars were then re-tempered at 500° and 550° C. for ¼, ½, 2, 8, 32, and 128 hr., respectively. Similar treatments were carried out at 575° C. for ¼, ½, 2, 8, 48, and 128 hr. The impact test results for all the treatments are shown in Figs. 10 and 11, and the transition temperatures and hardness values are given in Table II. The temper-embrittlement $\Delta\theta$ is plotted against the re-tempering time in Fig. 12*a* and against the logarithm of the time in Fig. 12*b*.

At 500° C. the temper-embrittlement in steel *K* continued to increase up to 128 hr., although the rate of embrittlement after about 32 hr. was extremely slow. At 550° C. the rate of embrittlement was initially greater than at 500° C., but it fell off very rapidly and the embrittlement reached a maximum of approximately 95° C. after about 6 hr. For re-tempering times greater than 6 hr., the embrittlement began to decrease and fell from 95° C. after 8 hr. to 74° C. after 128 hr. This meant that the embrittlement after 128 hr. was the same as that originally produced in the first 15 min. At 575° C. the embrittlement reached a maximum of about 72° C. after only ½ hr. and then decreased to 39° C. after 128 hr. The latter is considerably less than the value of 66° C. which was produced in the first 15 min.

Table II shows that, whereas during the treatments at 500° C. the hardness remained approximately constant, after the 128-hr. treatments at 550° and 575° C., it decreased by 14 and 23 V.D.H. numbers, respectively. Thus, for the specimens given the longer treatments at these temperatures, the transition temperatures may have been affected by the changes in the hardness, irrespective of any changes in the temper-embrittlement. To investigate the effect of decreasing the hardness on the transition temperature in steel *K*, a group of bars were oil-quenched from 850° C. and tempered at 650° C. for 5 hr. The results for this treatment are included in Fig. 13 and Table II, and indicate that a decrease of 27 V.D.H. numbers decreases the transition temperature by 8° C.

At 550° C. the transition temperature decreases by 21° C. between 8 hr. and 128 hr., whilst the hardness falls by only 10 V.D.H. numbers; at 575° C. the transition temperature decreases by 33° C. between ½ hr. and 128 hr. whilst the hardness falls by 18 V.D.H. numbers. Thus, provided that the change in the transition temperature produced by a given decrease in hardness is independent of the actual value of the transition temperature, it would appear that the softening at 550° and 575° C. is not sufficient to produce the observed decrease in the temper-embrittlement at these temperatures.

The etching attack produced by aqueous picric acid on the specimens of steel *K*, which had been re-tempered for various times at 500° C., is shown in Fig. 7. It is obvious that the intensity of the grain-boundary attack increases with increasing temper-embrittlement, although there is little difference between the specimens treated for 8, 32, and 128 hr. In the specimens given short treatments at 500° C. there is little indication of ferrite boundaries within the principal grain-boundary network, but in those treated for the longer times they are very evident and some of the ferrite boundaries appear to have been etched as deeply as the main network. Figure 7 shows that there is a slight grain-boundary etch present in the specimens which had not been intentionally temper-embrittled, and in steel *K* it was impossible to remove this attack from specimens

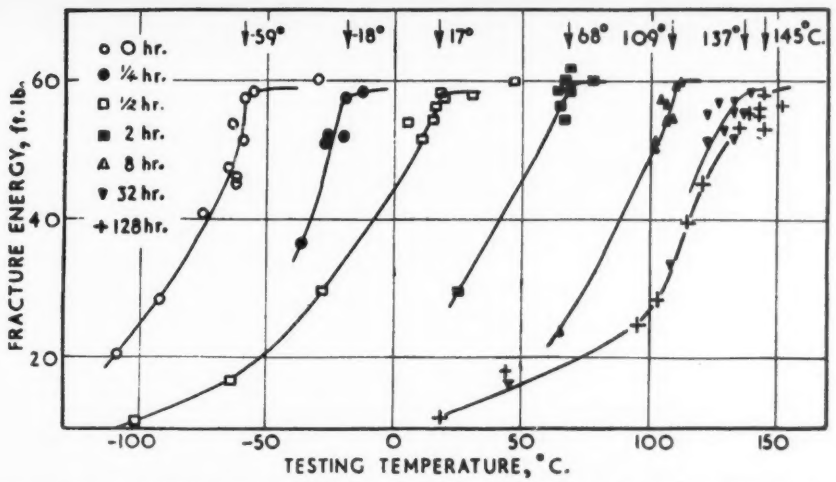


Fig. 10—Fracture-energy/testing-temperature curves for steel *K*, treated for various times at 500° C., after O.Q. 850° C. and T. 1 hr. at 650° C.

tempered at 650° C., even by rapid quenching of very small specimens. The extent and intensity of the etching attack produced by aqueous picric acid were less in the specimens treated at 550° and 575° C. than in those treated at 500° C., but otherwise they showed the same characteristics.

Electron microscope replicas were taken from specimens embrittled at 575° C. for 15 min. and 128 hr., respectively, but careful examination of the micrographs failed to reveal any difference between the etch produced in each by the aqueous picric acid. Figure 8 is an electron micrograph of an area in the specimen treated for 128 hr., and comparison of this with Fig. 6 shows that the appearance of the grain-boundary grooves is the same in each case, and that small particles occur in the boundaries in both specimens. There is no indication that the grain-boundary ' phase ' is spheroidizing or that precipitation is taking place in the boundaries.

Temper-Embrittlement at Temperatures above 575° C. —In view of the softening which took place in the treatments at 575° C. and the fact that the maximum

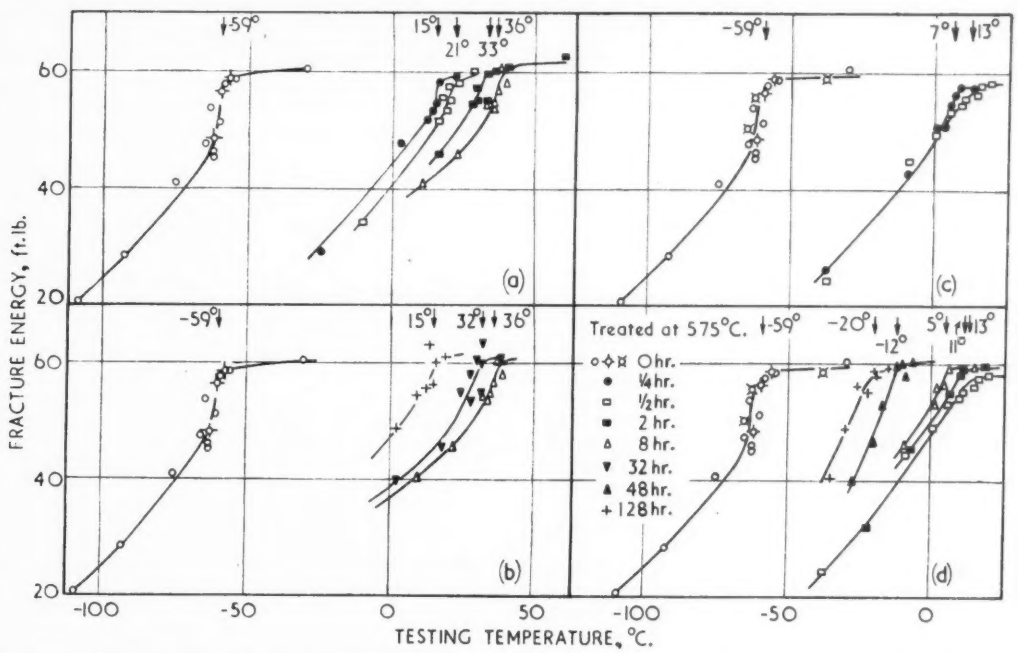


Fig. 11—Fracture-energy/testing-temperature curves for steel *K*, O.Q. 850° C., T. 1 hr. at 650° C., and treated for various times: (*a*) and (*b*) at 550° C., (*c*) and (*d*) at 575° C.

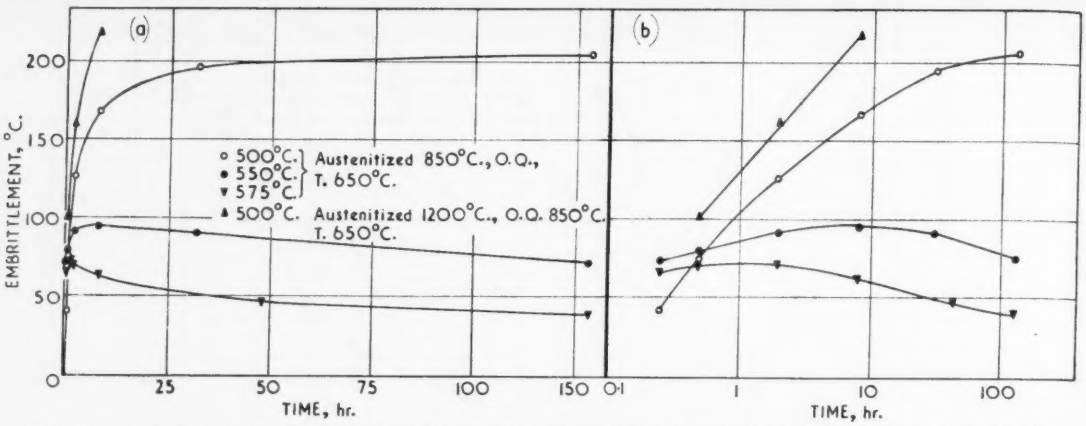


Fig. 12—Temper-embrittlement in steel *K* at various temperatures plotted against (*a*) time and (*b*) log time of treatment

temper-embrittlement was reached after 6 hr. at 550° C. and after ½ hr. at 575° C., it was decided not to carry out an extended series of treatments above 575° C. However, to see whether any embrittlement did occur above 575° C., two groups of bars were oil-quenched from 850° C. and tempered at 650° C. for 1 hr.; one group was embrittled at 600° C. for 15 min. and the other at 630° C. for 15 min. The results, given in Fig. 13 and Table II, show that in steel *K* temper-embrittlement can occur at and above 600° C. but that the magnitude of the effect is small. At 600° C. the embrittlement is about half that which occurred at 550° or 575° C. in the same time, and at 630° C. the embrittlement is only just detectable.

The specimens treated at 600° and 630° C. show a definite grain-boundary etch with aqueous picric acid, and the intensity of the attack is noticeably greater than that produced in the untreated specimens.

Double Temper-Embrittling Treatments—To investigate the effect of a previous treatment on the temper-embrittlement produced in a subsequent treatment at a different temperature, several double temper-embrittling treatments were carried out (*see* Table II).

In the first series, two groups of bars of steel *K*

Table III

DEVELOPMENT OF TEMPER-BRITTLENESS

Treatment*	Transition Temp., °C.	Embrittlement, °C.	V.D.H. (50 kg.)
(*a*) Effect of Microstructure			
Steel *J*			
O.Q., T. 1 hr. 650° C.	− 16		321
O.Q., T., emb. 24 hr. 525° C.	+ 60	76	312
Aust., 900° C., trans. ½ hr. 700° C.	+110		227
Trans. and emb. 24 hr. 525° C.	+127	17	223
Steel *K*			
Aust., trans. 620° C., T. ½ hr. 650° C.	+ 70		214
Aust., trans., T. and emb. 32 hr. at 500° C.	+121	51	210
Aust., trans. 350° C., T. ½ hr. 650° C.	+ 53		257
Aust., trans., T., and emb. 32 hr. 500° C.	+162	109	253
O.Q., T. 1 hr. 650° C.	− 59		280
O.Q., T., emb. 32 hr. 50 ° C.	+137	196	278
O.Q., T. 50 hr. 650° C.	− 52		223
O.Q., T., emb. 32 hr. 500° C.	+ 56	108	220
(*b*) Effect of Austenitizing Conditions			
Steel *K*			
Aust., ½ hr., 1200° C., then held 1 hr. 850° C. before O.Q., T. 1 hr. 650° C.	− 10		284
As above, emb. ½ hr. 500° C.	+ 91	101	280
As above, emb. 2 hr. 500° C.	+150	160	279
As above, emb. 8 hr. 500° C.	+209	219	278

* All the tempering and embrittling treatments were followed by water-quenching

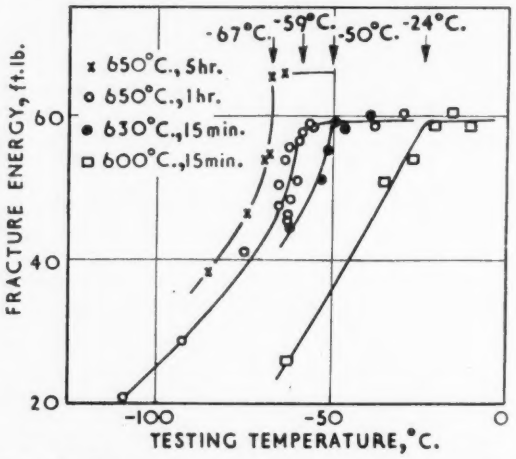


Fig. 13—Fracture-energy/testing-temperature curves for steel *K*, T. 1 hr. and 5 hr. at 650° C., and also reheated 15 min. at 600° C. and 630° C. after T. 1 hr. at 650° C.

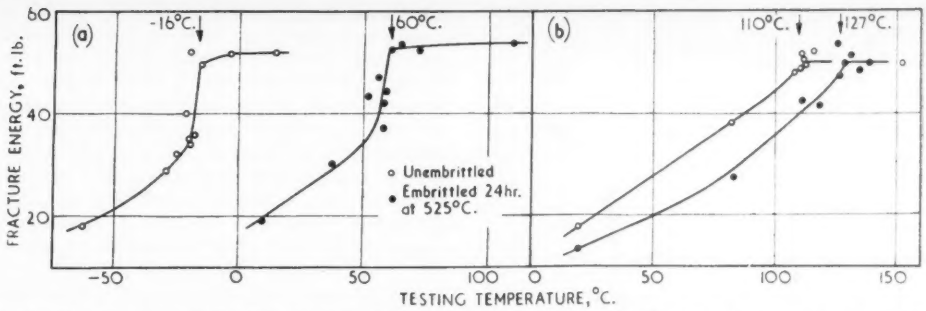


Fig. 14—Fracture-energy/testing-temperature curves for steel *J*: (*a*) Tempered martensite condition, (*b*) pearlitic condition

were oil-quenched from 850° C. and tempered at 650° C. for 1 hr. One group was treated at 500° C. for 8 hr., water-quenched, and then treated at 550° C. for 8 hr. The other group was given the reverse treatment, *i.e.*, treated for 8 hr. at 550° C., water-quenched, and treated for 8 hr. at 500° C. The impact results showed that the embrittlement produced in the double treatments was the same as that which would have occurred if the first of the treatments had been omitted, *i.e.*, the effect of 8 hr. at 500° C. followed by 8 hr. at 550° C. was the same as the effect of 8 hr. at 550° C.

In the second series, specimens that had been temper-embrittled for 128 hr. at 575° C. were reheated to 500° C. for ½ hr. and for 2 hr., respectively, and were then water-quenched. The impact test results showed that the specimens treated for 128 hr. at 575° C. and for ½ hr. at 500° C. had practically the same embrittlement as those treated for ½ hr. at 500° C., but that those treated for 128 hr. at 575° C. and 2 hr. at 500° C. showed less embrittlement than those treated for 2 hr. at 500° C. (*see* Table II).

Effect of Microstructure on the Development of Temper-Brittleness

It was decided to study the occurrence of temper-brittleness in pearlitic and bainitic microstructures and to compare its extent with that in the same steels in the tempered martensitic condition. Two steels, a 3% Cr steel (steel *J*) and a 3% Ni, 1% Cr steel (steel *K*), were used.

Pearlitic and Bainitic Microstructures—Bars of steel *J* were austenitized at 900° C., transferred to a furnace at 700° C. for 30 min., and then water-quenched. Microscopical examination showed that this treatment completely transformed the austenite to pearlite. Half the bars were treated at 525° C. for 24 hr. Further bars of steel *J* were oil-quenched from 900° C. and tempered at 650° C. for 1 hr.; half of these bars were also treated at 525° C. for 24 hr.

From the results obtained for the two microstructures, given in Fig. 14 and Table III*a*, it is clear that, although some temper-embrittlement has occurred in the pearlitic specimens, it is only about a quarter of that which occurred in the tempered martensitic specimens during the same treatment. However, there is a considerable difference in hardness between the two microstructures.

In steel *K*, preliminary experiments showed that a pearlitic microstructure could be obtained by transformation at 620° C. and a bainitic microstructure by transformation at 350° C. For each transformation the bars were austenitized at 850° C. for 1 hr., transferred to a furnace standing at the transformation temperature, held for 24 hr., and then water-quenched. At both 620° and 350° C. the austenite was completely transformed after 24 hr. and no pearlite was found in the microstructure of the specimens transformed at 350° C. To remove any temper-brittleness which might have occurred during the transformation, all the bars were tempered at 650° C. for ½ hr. and

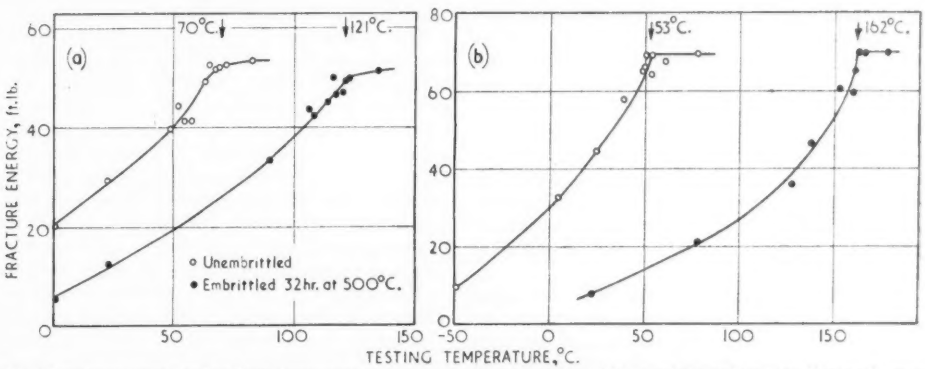


Fig. 15—Fracture-energy/testing-temperature curves for steel *K*: (*a*) Transformed to pearlite and temper-embrittled, (*b*) transformed to bainite and temper-embrittled

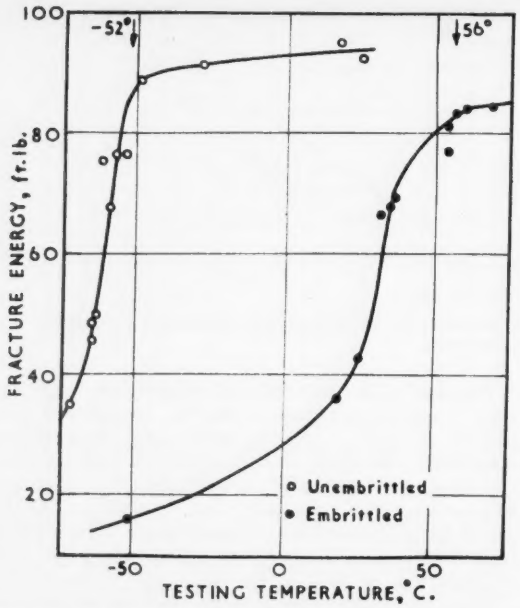


Fig. 16—Fracture-energy/testing-temperature curves for steel *K*, T. 50 hr. at 650° C. and temper-embrittled for 32 hr. at 500° C.

water-quenched. Half the bars from each group were subsequently treated at 500° C. for 32 hr. The notched-bar impact test results for the pearlitic specimens are shown in Fig. 15*a* and those for the bainitic specimens in Fig. 15*b*, whilst the respective transition temperatures and hardness values are given in Table III*a*. The corresponding results for steel *K* in the quenched and tempered condition had already been obtained in the time and temperature experiments reported in the previous section (Fig. 10) and are included in Table III for comparison.

The results show that temper-brittleness can occur in pearlitic and bainitic microstructures as well as in tempered martensite, and in steel *K*, for the particular transformation and embrittlement treatments used, the embrittlement $\Delta\theta$ is 196° C. for tempered martensite, 110° C. for bainite, and 51° C. for pearlite.

The pearlitic and bainitic specimens were etched with aqueous picric acid, and in both microstructures the reagent produced a heavy grain-boundary attack in the temper-embrittled specimens and a slight grain-boundary etch in the unembrittled ones. For the same time of etching, the grain-boundary attack was not so pronounced as in the tempered martensitic specimens, but the general characteristics of the etch were the same. An electron micrograph of the etch produced in the temper-embrittled pearlitic specimens is shown in Fig. 9. Similar results were obtained with the aqueous picric acid solution for the pearlitic specimens of steel *J*.

Effect of Prolonged Tempering at 650° *C.*—To study the effect of a prolonged tempering treatment at 650° C. on the subsequent development of temper-brittleness, bars of steel *K* were oil-quenched from 850° C. and tempered at 650° C. for 50 hr.; half the bars were then treated at 500° C. for 32 hr. The impact test results are given in Fig. 16 and the hardness and transition temperatures are included in Table III*a*. Comparison of these results with those obtained for steel *K* quenched and tempered at 650° C. for 1 hr. (*see* Fig. 10 and Table III*a*) shows that, as expected, the hardness has been considerably reduced and the maximum fracture energy has correspondingly increased, but that the embrittlement produced by the 32-hr. treatment at 500° C. has been reduced from 196° to 108° C. The transition temperature of the steel without temper-brittleness has been raised slightly from — 59° to — 52° C., but that of the temper-embrittled steel has been reduced from 137° to 56° C.

Effect of Austenitizing Conditions on the Development of Temper-Brittleness

Very little data are available on the effect of the austenitizing time and temperature upon the subsequent development of temper-brittleness, and although previous investigators seem to agree that increasing the grain size of a steel increases the susceptibility to temper-embrittlement,[4, 9, 10] there is no information on the magnitude of such an effect.

Effect of Austenitizing Time at 850° *C.*—Groups of bars of steel *K* were austenitized for ½ hr. and 10 hr. at 850° C. and then tempered at 650° C. for 1 hr. Half of each group was then treated at 500° C. for 8 hr. The impact results for these treatments lay on the corresponding curves obtained for specimens of steel *K*, which had been austenitized for 1 hr. (the standard austenitizing time used in the present work), and so the austenitizing time at 850° C. does not appear to affect the subsequent temper-embrittlement.

Effect of Austenitizing at 1200° *C. on Subsequent Development of Temper-Brittleness*—Bars of steel *K* were austenitized at 1200° C. for ½ hr. and were then transferred for 1 hr. to a furnace at 850° C., after which they were oil-quenched. All the bars were tempered at 650° C. for 1 hr., and groups were then treated at 500° C. for ½, 2, and 8 hr., respectively.

To investigate whether the quenching temperature, as distinct from the austenitizing temperature, had any effect on the subsequent development of temper-brittleness, two groups of bars of steel *K* were oil-quenched directly from 1200° C. They were then tempered at 650° C. for 1 hr. and one group was treated at 500° C. for 8 hr.

The results for those bars quenched from 850° C. are given in Fig. 17 and Table III*b*. The temper-embrittlement which occurred in this coarse-grained steel has been plotted against the time of treatment and the logarithm of the time in Fig. 12, and the curves show that both the embrittlement after a given time and the rate of embrittlement are much greater than for the same steel in the fine-grained condition. However, the impact properties of the coarse-grained are appreciably different from those of the fine-grained material (*cf.* Figs. 10 and 17).

The specimens of the steel quenched directly from 1200° C. and tempered at 650° C. had a slightly lower transition temperature than those quenched from 850° C. and similarly tempered, but the fracture-energy curves for the two groups of temper-embrittled

specimens were the same. In this case the quenching temperature, as distinct from the austenitizing temperature, appeared to have no effect upon the subsequent development of temper-brittleness.

The temper-embrittled specimens all showed a pronounced grain-boundary etch with aqueous picric acid, the intensity of the attack increasing with the retempering time. The brittle fracture in the specimens which had not been temper-embrittled included both intergranular and cleavage facets, whilst that in the temper-embrittled specimens was completely intergranular.

Effect of Titanium and Aluminium on the Occurrence of Temper-Brittleness

If, as several investigators have suggested, temper-brittleness is due to the precipitation of a nitride from solution in the ferrite,[11, 12, 13] then the addition of either titanium or zirconium, which form nitrides with a considerably greater stability than those of Fe, Cr, and Mn, should decrease the temper-embrittlement in a susceptible steel.

In the present work, a series of Ni–Cr–Ti steels were made, of which *B*, *C*, and *D*, containing 0, 0·47, and 1·5% of Ti, respectively, were examined. When the alloys were prepared, the only source of Ti was ferro-titanium, which contained appreciable amounts of Al; thus to investigate the separate effect of this element a Ni–Cr–Al steel *H* was made at the same time. To study the effect of Ti in the absence of Ni and Cr, steel *G* was also prepared.

The analyses of these steels (Table I) show that the total nitrogen content of the steels containing Ti was appreciably below that of the Ni–Cr steel, but that the Ni–Cr–Al steel appeared to have a much greater nitrogen content. However, whilst the method used to determine the total nitrogen contents of the steels would certainly involve the decomposition of any aluminium nitride present, it would most probably fail to break down the refractory titanium nitride. Thus, the nitrogen content reported for the steels containing Ti would be the nitrogen present either in solution in the ferrite or combined as iron, chromium, and aluminium nitrides, and would not include that present as titanium nitride. In the Ni–Cr–Al steel the increased nitrogen was presumably picked up on casting after the addition of the Al.

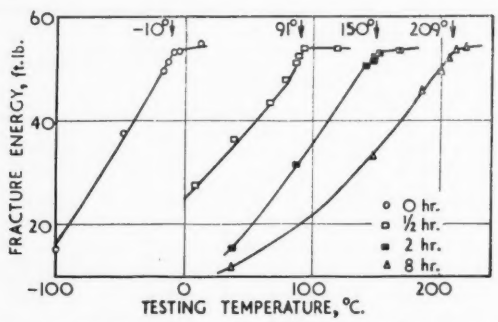
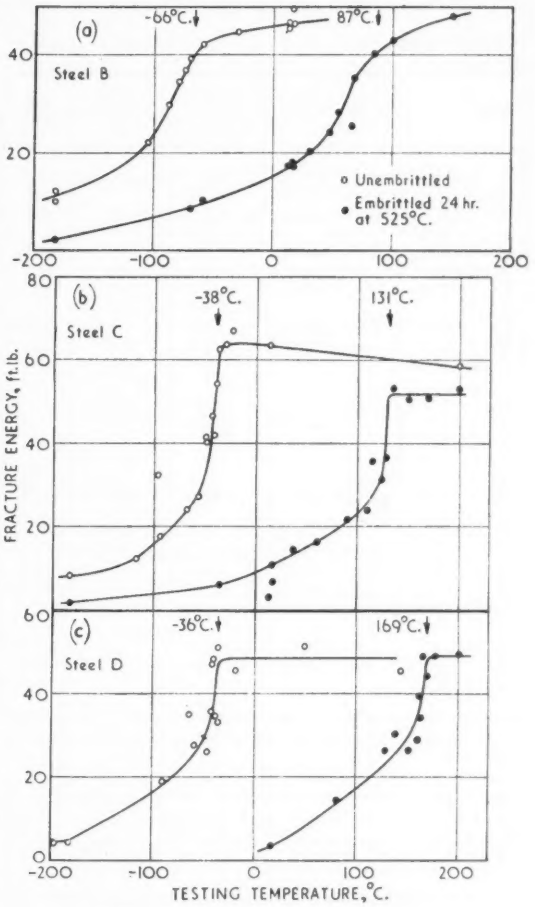


Fig. 17—Fracture-energy/testing-temperature curves for steel *K*, austenitized 1200° C., O.Q. 850° C., T. 1 hr. at 650° C., and treated for various times at 500° C.



(*a*) Steel *B*, O.Q. 850° C., T. 1 hr. at 650° C.
(*b*) Steel *C*, O.Q. 850° C., T. 1 hr. at 650° C.
(*c*) Steel *D*, O.Q. 900° C., T. 1 hr. at 650° C.

Fig. 18—Fracture-energy/testing-temperature curves

Effect of Titanium Additions to a Ni–Cr Steel—Bars from steels *B* and *C* were oil-quenched from 850° C. and tempered at 650° C. for 1 hr.; half the bars from each group were then treated for 24 hr. at 525° C. Steel *D* was oil-quenched from 900° C. and then tempered and treated as for steels *B* and *C*.

The fracture-energy/testing-temperature curves obtained for the three steels are given in Fig. 18 and the transition temperatures and hardness values in Table IV*a*. From these results it is obvious that the addition of Ti to a susceptible Ni–Cr steel does not prevent temper-embrittlement. With the particular heat-treatments used, the Ni–Cr–Ti steels actually showed a greater susceptibility to temper-embrittlement than did the Ni–Cr base alloy. The pronounced scatter in the results for the steels containing Ti is due to the presence of bands of titanium nitride and titanium carbide particles in the specimens, which give rise to torn fractures and correspondingly high fracture energies.

All the temper-embrittled steels gave a grain-

Table IV

OCCURRENCE OF TEMPER-BRITTLENESS

Steel	Treatment*	Trans. Temp., °C.	Embrittlement, °C.	V.D.H. (50 kg.)
	(*a*) Effect of Titanium and Aluminium			
B	O.Q., T. 1 hr. 650° C.	− 66		288
B	O.Q., T., emb. 24 hr. 525° C.	+ 87	153	280
C	O.Q., T. 1 hr. 650° C.	− 38		284
C	O.Q., T., emb. 24 hr. 525° C.	+131	169	276
D	O.Q., T. 1 hr. 650°C.	− 36		270
D	O.Q., T., emb. 24 hr. 525° C.	+169	205	265
G	O.Q., T. 1 hr. 650° C.	+ 90		
G	O.Q., T., emb. 24 hr. 525° C.	+125	35	
H	O.Q., T. 1 hr. 650° C.	− 27		232
H	O.Q., T., emb. 24 hr. 525° C.	+178	205	250
	(*b*) Nickel and Nickel-Phosphorus Steels			
T	W.Q., T. 1 hr. 650° C.	− 34		207
T	W.Q., T., emb. 24 hr. 500° C.	− 27	7	204
U	W.Q., T. 1 hr. 650° C.	+ 12		236
U	W.Q., T., emb. 24 hr. 500° C.	+ 96	84	232

*All the tempering and embrittling treatments were followed by water-quenching

boundary etch with aqueous picric acid, although in the Ni–Cr–Ti steels the etch developed very slowly. The brittle fractures of all the three steels in the temper-embrittled condition were intergranular.

Iron–Carbon–Titanium Alloy—The original intention was to produce a Ti steel that could be quenched to martensite and then tempered and embrittled as for the Ni–Cr–Ti alloys, but the Ti losses in melting had been less than expected and, as a result, the composition of the alloy fell within the (α + TiC) region of the Fe–Ti–C ternary system, and no change point was found in dilatometer specimens heated to 1100° C. In spite of this, it was decided to heat-treat the alloy in the same way as the Ni–Cr–Ti steels to see whether temper-embrittlement would develop in a ferritic alloy.

Bars of the steel (*G*) were oil-quenched from 900° C. and tempered for 1 hr. at 650° C.; half of the bars were then treated at 525° C. for 24 hr. The results are given in Fig. 19*a* and Table IV*a*. The scatter in the impact results is again due to the pronounced banding exhibited by the titanium carbide and nitride particles.

In spite of the scatter in the results, it is clear that the steel has been embrittled by the treatments at 525° C., although the increase in the transition temperature is only about 35° C. However, no grain-boundary etching attack could be produced with aqueous picric acid in microsections from either of the groups of specimens, and the brittle fracture of specimens from both groups appeared to be a cleavage fracture, with no evidence of intergranular facets. Thus, although the transition temperature has been displaced by the treatment at 525° C., it must be concluded that this embrittlement is not of the same type as that in the Ni–Cr–Ti steels and is not temper-brittleness in the sense in which this term is used in the present paper.

Effect of Aluminium—Bars of steel *H* were oil-quenched from 850° C. and tempered at 650° C. for 1 hr.; half of them were then treated at 525° C. for 24 hr. The impact results are given in Fig. 19*b* and Table IV*a*. The presence of Al has not prevented temper-embrittlement. Large numbers of blue-grey non-metallic inclusions (presumably aluminium nitride) were present in the microsections, and the fractures of the impact specimens were badly torn, although the resulting scatter in the fracture-energy results was not appreciable.

The temper-embrittled specimens of steel *H* showed a distinct grain-boundary etch with aqueous picric acid, and the brittle fractures were completely intergranular.

Table V

STEEL *K*: RESULTS OF LOW-TEMPERATURE TENSILE TESTS

Condition	Testing Temp., °C.	Yield Stress	Nominal Max. Stress	True Breaking Stress	Reduction in Area, %	Fracture
		Tons/sq. in.				
O.Q., T.	+ 20	51·5	64·0	...	65·5	Ductile
O.Q., T., emb. 32 hr. 500° C.	+ 20	51·5	63·5	...	62·5	Ductile
O.Q., T.	−196	73·8	96·5	...	57·0	Ductile
O.Q., T., emb. 32 hr. 500° C.	−196	75·0	84·1	126·0	35·0	Brittle
Bainitic	+ 20	46·8	59·2	...	59·4	Ductile
Bainitic	−196	71·1	86·1	129·5	36·2	Brittle
Bainitic, emb. 3 hr. 500° C.	−196	70·3	83·0	121·1	35·6	Brittle
Bainitic, emb. 32 hr. 500° C.	−196	71·5	85·4	96·7	11·6	
Pearlitic	−196	57·1	72·0	105·4	33·7	Brittle
Pearlitic, emb. 32 hr. 500° C.	−196	57·3	...	79·6	9·8	Brittle

Occurrence of Temper-Brittleness in a Nickel Steel

Nickel steels are not usually considered to be temper-brittle,[7, 11, 14] but only a little reliable information on this point is available, and it was decided to investigate whether temper-embrittlement could in fact be produced in a nickel steel.

A 3% Ni steel *T* was quenched in brine from 900° C. and tempered at 650° C. for 1 hr.; half the bars were then treated at 500° C. for 24 hr. The results, given in Fig. 19*c* and Table IV*b*, show that a very slight embrittlement has been produced by the treatment at 500° C., the actual displacement of the transition temperature being only 7° C.

A slight but definite grain-boundary etch was produced by the aqueous picric acid in both the treated and the untreated specimens, but the intensity of the attack was undoubtedly greater in the embrittled specimens. The brittle fracture in the temper-embrittled specimens comprised both intergranular and cleavage facets.

Occurrence of Temper-Brittleness in a Nickel-Phosphorus Steel

Several investigators have shown that if the phosphorus content of a temper-brittle steel is increased, the temper-embrittlement is also increased,[1, 15, 16, 17] but no data are available to show whether phosphorus alone can produce temper-brittleness. It was therefore decided to prepare a nickel steel similar to steel *T*, but with approximately 0·1% of phosphorus, and to investigate its susceptibility to temper-brittleness.

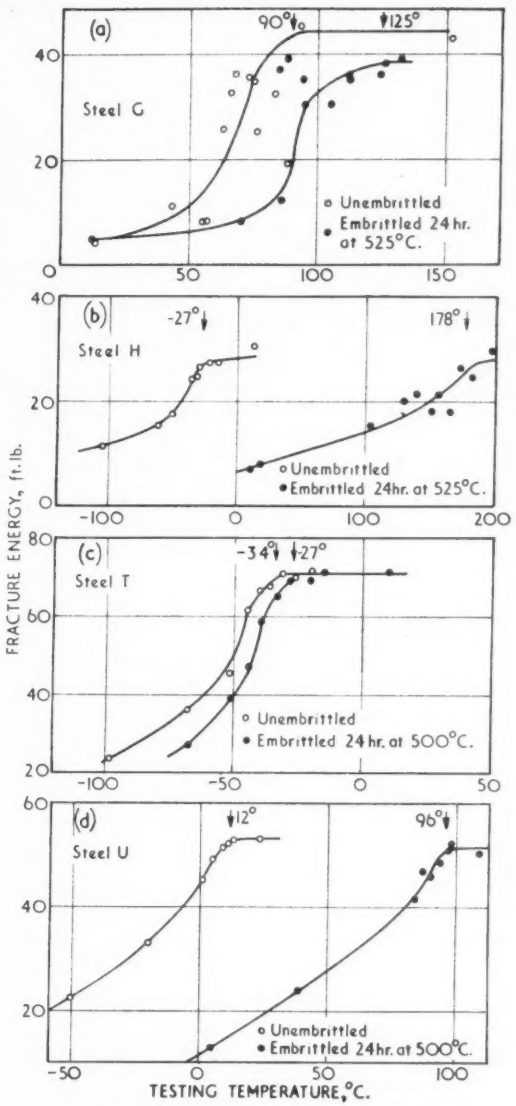
This steel (*U*) was water-quenched from 900° C. and tempered at 650° C. for 1 hr.; half the bars were then treated at 500° C. for 24 hr. The results, given in Fig. 19*d* and Table IV*b*, show that a considerable embrittlement has been produced by the treatment at 500° C. and that the displacement of the transition temperature has been increased to 84° C., compared with 7° C. for steel *T*. Also, compared with steel *T*, the properties of steel *U* in the absence of temper-brittleness were adversely affected by the increased phosphorus; the maximum fracture energy was decreased from 72 to 52 ft.lb. and the transition temperature was raised from — 34° to + 12° C.

A definite grain-boundary etch was produced in the temper-embrittled specimens, and the brittle fracture was intergranular.

Low-Temperature Tensile Tests

It is generally accepted that temper-brittleness has no effect on normal room-temperature mechanical properties.[4, 11, 14, 18] However, after considering the theory of the notched-bar impact test and the effect produced by temper-brittleness on the notched-bar impact results of a susceptible steel, it was concluded that temper-brittleness must lower the brittle-fracture strength of the steel, and hence the effect should be apparent in tensile tests carried out at sub-zero temperatures. Low-temperature tensile tests were therefore carried out on steel *K*, which had been shown to be extremely susceptible to temper-brittleness.

The ductility of steel *K* in the tempered martensitic condition was such that even at — 196° C. it did not show a brittle fracture but failed in a ductile manner.



(*a*) Steel *G*, O.Q. 900° C., T. 1 hr. at 650° C.
(*b*) Steel *H*, O.Q. 850° C., T. 1 hr. at 650° C.
(*c*) Steel *T*, brine-quenched 900° C., T. 1 hr. at 650° C.
(*d*) Steel *U*, W.Q. 900° C., T. 1 hr. at 650° C.

Fig. 19—Fracture-energy/testing-temperature curves

The reduction in area was 57%, compared with 65·6% at room temperature (*see* Table V). After 32 hr. at 500° C. the reduction in area was reduced to 35% and the maximum stress from 96·5 to 84·1 tons/sq. in. These results indicate that the temper-brittleness had reduced the brittle-fracture strength, but they are not conclusive.

As tensile tests were not possible below — 196° C., the effect of temper-brittleness upon the tensile properties of steel *K* in the bainitic and pearlitic conditions was examined. The steel was transformed and heat-treated as for the impact tests, except that the bainitic specimens were treated for 3 hr. and

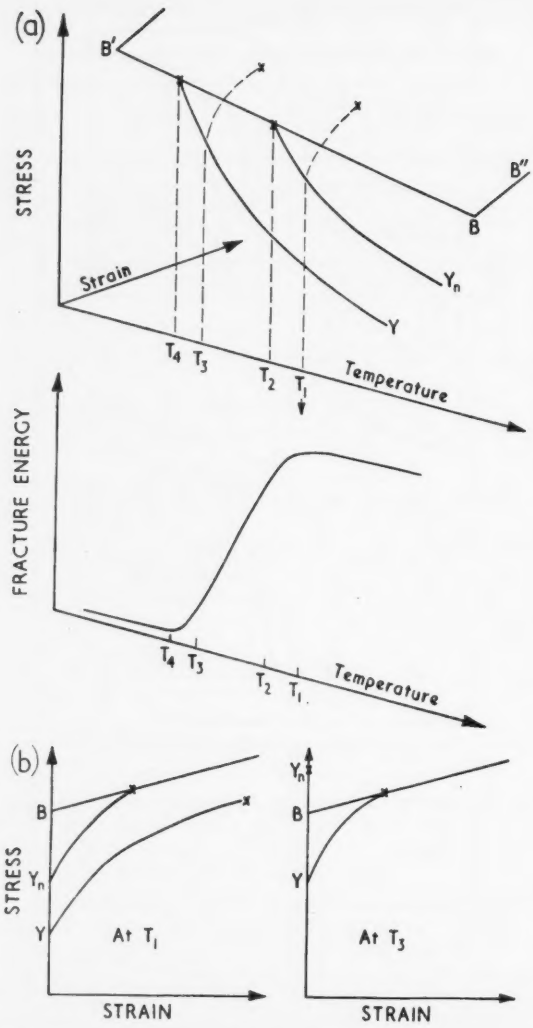


Fig. 20—Diagrammatic derivation of fracture-energy/testing-temperature curve: (*a*) Effect of temperature and strain upon yield stress and brittle-fracture strength; (*b*) sections through diagram (*a*) at temperatures T_1 and T_3

32 hr. at 500° C., and the pearlitic specimens for 32 hr. at 500° C. The results (Table V) show that temper-brittleness reduces the brittle-fracture strength of both microstructures without affecting the yield stress or, in the bainitic specimens, the maximum stress. Both microstructures showed a brittle fracture in the absence of temper-brittleness although appreciable plastic deformation took place in each case before the failure occurred.

With the bainitic microstructure the stress at fracture fell from 129·5 to 121·1 tons/sq. in. after 3 hr. at 500° C. and then to 96·7 tons/sq. in. after 32 hr. at 500° C. These figures are calculated from the final breaking load and the final cross-sectional area, but the untreated specimen and the specimen treated for 3 hr. had necked down slightly before fracture occurred. This meant that the stress distribution was not uniform and the values of the stress at fracture are only the mean values. With the pearlitic microstructures the untreated specimen failed just after necking and the temper-embrittled specimen before the maximum load. The stress at fracture was reduced from 105·4 to 79·6 tons/sq. in. by the 32-hr. treatment at 500° C.

Whilst the tempered martensitic and the bainitic microstructures did not show a discontinuous yield at room temperature, they did so at — 196° C.

THEORY OF THE NOTCHED-BAR IMPACT TEST

The notched-bar impact test has been examined in some detail and it is considered that the relationship between it and the fundamental flow and fracture properties of a steel can be represented as in Fig. 20. It is assumed that brittle fracture occurs when the greatest normal tensile stress reaches a certain critical value, and that the brittle-fracture strength increases with increasing strain.

As a result of the tri-axial stress system set up in a notched-bar impact specimen, and also as a result of the high strain rate, the yield stress of the metal in the bar is raised above that measured in the normal tensile test. The yield stress will not be constant across the bar but will be a maximum at some point just below the centre of the notch and will decrease towards the sides of the bar. Applying Orowan's theory of notch-brittleness[19, 20] to the present problem, the maximum value of the yield stress will be Y_n, where $Y_n/Y \simeq 2$.

In Fig. 20*a*, curve Y shows the variation of the yield stress with temperature, and curve Y_n the corresponding variation of the maximum yield stress in the notched-bar impact specimen. The dependence of the brittle-fracture stress on temperature is given by line B, and the variation of the brittle-fracture stress with both temperature and strain is given by the surface $B\,B'B''$. The surfaces showing the effect of strain on Y and Y_n have been omitted for clarity. The broken lines in Fig. 20*a* show some of the stress/strain curves for the materials with yield stress Y_n and Y at temperatures T_1, T_2, T_3, T_4. The curves at T_1 and T_3 terminate on the plane $B\,B'B''$. Sections through the diagram in Fig. 20*a* are given in Fig. 20*b* for temperatures T_1 and T_3, and here the intersections of the Y/strain/temperature surface and the Y_n/strain/temperature surface are shown. They correspond, of course, to the true-stress/true-strain curves for the materials with yield stresses Y and Y_n at the temperatures T_1 and T_3.

From Fig. 20 it is clear that, allowing for the effect of strain upon both the yield stress and the fracture strength, brittle fracture will first appear in the notched-bar impact specimens at a temperature T_1. Brittle fracture will not occur over the whole specimen until a temperature T_3. Between T_1 and T_3 part of the bar will show a brittle fracture and part a ductile fracture. The shape of the actual fracture-energy/testing-temperature curve will depend on the relationships between Y, Y_n, B, and their variations with strain and temperature. The transition temperature, as defined in this work, obviously corresponds to the temperature T_1 in Fig. 20.

The effect of temper-brittleness upon the notched-bar impact properties of a susceptible steel is to increase the temperature at which brittle fracture first appears in the specimens, to decrease the energy required for complete brittle fracture, and to leave the energy required for ductile fracture unchanged. These effects can be accounted for in terms of the diagrams in Fig. 20, if the result of temper-brittleness is to decrease the brittle fracture strength from B to a value B_i without affecting the flow properties of the steel. This is shown in Fig. 21, where the effect of strain has been omitted for clarity.

If the theory outlined above is correct, then it is obvious that the displacement of the transition temperature produced by a given decrease in the brittle-fracture strength depends on the relative slopes of the Y_n/temperature curve and the B_i/temperature curve, and this is indicated in Fig. 21. This must be borne in mind when comparing the temper-embrittlement in materials where the mechanical properties are widely different.

The foregoing experimental results indicate that curve B in Fig. 21 refers to cleavage fracture and curve B_i to intergranular fracture. It may be postulated that for all steels there exists an intergranular-fracture stress, but in most cases it is greater than B and so cleavage fracture is generally observed. Embrittlement will be observed when, as the result of temper-brittleness or any other grain-boundary phenomenon, B_i is brought below B.

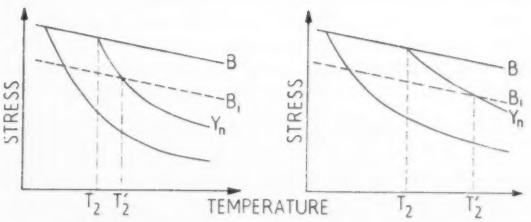


Fig. 21—Diagrams showing the effect of a given decrease in the brittle-fracture strength produced by temper-brittleness, on the displacement of the transition temperatures in two steels with different yield-stress/temperature curves

DISCUSSION

Metallographic Results

The results of the thermal etching experiment show that the main etching attack produced by aqueous picric acid in the temper-embrittled steels occurs along what were the austenite grain boundaries, but that some of the ferrite boundaries are also attacked. Although these results were obtained for a steel which had transformed to bainite, there is no reason to suppose that the conclusions drawn from them do not also apply to steels which have been quenched to martensite before tempering and temper-embrittling. This is borne out by the electron micrographs, Figs. 6 and 8, from the quenched, tempered, and temper-embrittled specimens of steel K, which clearly show that the relationship between the two sets of boundaries produced by the aqueous picric acid is identical with that between austenite and ferrite boundaries.

Although the electron micrographs show that the aqueous picric acid had produced grooves along the various boundaries in both the tempered martensite and the pearlite, they are of little help in identifying the cause of the attack. They do show, however, that the carbide particles in both microstructures stand proud from the matrix and have not been preferentially attacked by the reagent. Aqueous picric acid has been used on a number of alloys, but apart from the grooves produced in the temper-brittle steels, its action appeared to be very similar to that of normal picral; in Armco iron, for example, the effect of aqueous picric acid was the same as that of picral in producing grain-boundary grooves or grain-boundary ridges. The only conclusion that can be drawn is that the grooves do not appear to be caused by any of the phases or inclusions normally occurring in steels.

The results obtained for the effect of aqueous picric acid on the various steels show that the reagent gives an indication of the extent of the temper-embrittlement, but that the etching conditions must be carefully standardized. Comparison of the etching attack on steels of different compositions can be very misleading as an indication of the extent of the relative embrittlements.

As an incidental observation, the micrograph of the thermal etch, Fig. 4*a*, reveals a number of surface-relief effects which must have been produced during the transformation of the austenite to bainite. In some areas the surface is roughened, but in others a number of straight lines appear across what were the austenite grains. The most prominent of these striations seem to coincide with ferrite areas (*see* Fig. 4), but many of the other less prominent ones do not. The appearance of the surface markings, as a whole, suggests that in some way the bainite has formed from the austenite by a shear process.

Effect of Time and Temperature

The decrease in the temper-embrittlement which occurred in steel K after long-time treatments at 550° and 575° C., does not appear to be due either to the softening which accompanies such treatments or to the spheroidization of the grain-boundary ' phase.' A possible explanation is that it is the result of the diffusion of Cr from the matrix into the carbide particles, thus reducing the Cr content of the ferrite. Bowers[21] has established that such an enrichment of the carbide in Cr occurs as tempering proceeds. If this is correct, it would be expected that the decrease in the temper-embrittlement with increasing tempering time would not be so noticeable in steels where the temper-brittleness was caused by Mn, Mo, or W, as a result of the slower rates of diffusion of these elements. However, no data are yet available on this point. Although the embrittlement continues to increase at 500° C. at tempering times up to 128 hr., it would be expected to decrease after longer times.

It appears from the etching attack in steel K that temper-brittleness can occur at temperatures up to the Ae_1 temperature (approximately 665° C. in this steel), although the actual embrittlement above 600° C. is very small (*see* Fig. 13 and Table II). From the discussion on the notched-bar impact test it will be apparent that, as the intercrystalline cohesion is originally greater than the cleavage strength, a

certain amount of temper-brittleness must occur before any actual embrittlement results. It is presumably this initial temper-brittleness that is responsible for the etching attack in specimens tempered above 625° C. which show hardly any embrittlement.

Effect of Microstructure and Grain Size

The differences in the degree of temper-embrittlement shown by the different microstructures appear to be partly due to the different relationships between the flow and fracture stresses and partly to differences in the distribution of alloying elements between the ferrite and carbide. In particular, the Cr content of the ferrite will be much lower in the pearlitic than in the tempered martensitic microstructure.

The impact results for steel *K*, after the long-time tempering at 650° C. and in the absence of temper-embrittlement, when compared with the corresponding results for the isothermally transformed structures, illustrate the well-known fact that the notched-bar impact properties of tempered martensite are considerably better than those of any other structure of comparable hardness. The respective transition temperatures and values for the maximum fracture energy are — 52° C. and 90 ft.lb. for tempered martensite, + 50° C. and 70 ft.lb. for bainite, and + 70° C. and 52 ft.lb. for pearlite.

The difference in temper-embrittlement observed between the specimens of steel *K* tempered for 1 hr. at 650° C. and those tempered for 50 hr. may be due to the diffusion of Cr from the ferrite into the carbide or to the change in the fracture and flow properties of the steel, as a result of softening and of ferrite grain growth. In the absence of a complete series of mechanical tests it is not possible to be certain which of these two changes is responsible.

Although Hollomon[11] has suggested that if temper-brittleness is a precipitation phenomenon, increasing the grain size should decrease the rate of temper-embrittlement, the results for steel *K* show that both the embrittlement produced in a given time and the rate of embrittlement are increased for the large grain size. Hollomon's argument was based on the increased distances over which diffusion would have to occur, but it appears that any such effect is offset by the decrease in the interfacial area of the austenite. The mechanical properties of the steel are considerably affected by the increase in the austenite grain size (*cf.* Figs. 10 and 17) and the greater embrittlement may be partly the result of this and not entirely due to an increase in the amount of temper-brittleness.

Theory of Temper-Brittleness

A number of theories have been advanced to explain the phenomenon of temper-brittleness,[2] although none has met with universal agreement. The most widely held theory is that temper-brittleness is due to the precipitation of one or more compounds from solution in the ferrite, *e.g.*, carbides, nitrides, phosphides, and oxides. However, the present results do not support the precipitation theory in several important respects.

The grain-boundary etch produced by aqueous picric acid in embrittled steels does not show a precipitated phase, but a series of grooves. These grooves are not made up of a series of pits from which particles have been etched out, but appear to be definite channels, the depth and width of which vary from place to place. The width and depth of the grooves are increased by increasing the etching time, a result which would not be expected if the attack were due to a film of precipitate which possessed definite boundaries. It is difficult to imagine the nature of a phase which is preferentially removed from the matrix by aqueous picric acid and yet is not attacked by any of the normal reagents. The aqueous picric acid does not attack iron and chromium carbides, as shown by Figs. 6, 8, and 9, and it would be expected that the presence of such compounds as nitrides, phosphides, and oxides would be revealed by the normal reagents.

As no precipitate has been found in the temper-embrittled specimens, it is reasonable to consider an alternative explanation of the etching attack produced by aqueous picric acid. The appearance of the grooves and particularly their development with etching time suggest that the attack might be due to changes in the composition of the matrix at the various boundaries, and of the two possibilities of depletion or enrichment of solute atoms the latter is the more likely. Apart from the fact that on theoretical grounds it would be expected that solute atoms would tend to diffuse to the grain boundaries,[22–24] there is an increasing body of experimental evidence to show that grain-boundary segregation of impurity atoms occurs in a large number of alloy systems.[25–30]

The fact that in temper-embrittled steels the main grain-boundary etching attack and the brittle fracture follow the prior austenite grain boundaries means that some of the changes which give rise to temper-brittleness must take place in the austenite and persist after the transformation. As temper-brittleness can be largely avoided by quenching from 650° C., it follows that the embrittlement is also associated with changes occurring in the ferrite. It is therefore suggested that certain solute atoms segregate to the grain boundaries, both in the austenite and in the ferrite, and the combination of these two processes gives rise to temper-brittleness. Essentially, this suggestion is a modification of that made by McLean and Northcott[7] to explain temper-brittleness.

It is not easy, however, to decide which elements are responsible for temper-brittleness. Those which seem to give rise to temper-brittleness when added to a plain carbon steel include Cr, Mn, W, and Mo, whilst those which have no effect include Al and S.[2] The present investigation indicates that phosphorus is able to produce temper-embrittlement in the presence of Ni and that plain Ni steels are only very slightly susceptible. Also, Ti does not prevent temper-embrittlement and this would appear to eliminate nitrogen from the list of possible elements. Therefore, temper-brittleness may be due to the segregation of atoms of Cr, Mn, Mo, W, and possibly phosphorus to the austenite grain boundaries, combined with the segregation of carbon and possibly phosphorus to the ferrite grain boundaries. The segregation of carbon in the ferrite will increase as the temperature decreases, and thus the lower the temperature the greater is the maximum possible temper-embrittlement. However, the controlling factor at low temperatures will be the rate of diffusion, and hence for a

given time the temper-embrittlement will show a maximum at some temperature below Ae_1 and will decrease at both lower and higher temperatures.

The absence of temper-embrittlement in specimens quenched from above 630° C. can be accounted for by a decrease in the grain-boundary segregation of carbon atoms at such temperatures. This decreased segregation will result both from the increased thermal mobility of the carbon atoms and, as the ferrite lattice expands, from a decrease in the lattice strain produced by the presence of the carbon.

The result of the segregations is to reduce the cohesion across what were the austenite interfaces to a value below that of the ferrite cleavage planes, and thus to lower the brittle-fracture stress and change the mode of brittle fracture. It has been shown in the previous section how the lowering of the brittle-fracture strength produces the observed changes in the mechanical properties of the susceptible steels.

CONCLUSIONS

(1) A grain-boundary etch can be produced in temper-brittle steels by a simple aqueous solution of picric acid.

(2) The etching reagent produces grooves at both the austenite and ferrite boundaries and does not reveal a grain-boundary precipitate.

(3) Brittle fracture occurs in tempered specimens by cleavage across the ferrite grains, and in temper-embrittled specimens by failure along the previously existing austenite boundaries.

(4) Temper-brittleness reduces the brittle-fracture stress of a susceptible steel without affecting the yield stress or the strain-hardening.

(5) The temper-embrittlement at 550° and 575° C. in a Ni–Cr steel reached a maximum and then decreased as the tempering time increased.

(6) Temper-brittleness takes place in a Ni–Cr steel up to the Ae_1 temperature.

(7) Prolonged tempering of a Ni–Cr steel at 650° C. reduced the subsequent temper-embrittlement.

(8) Temper-brittleness occurs in isothermally transformed microstructures although the extent of the temper-embrittlement is less than in the corresponding tempered martensitic structure.

(9) Increasing the grain size of a susceptible steel increases both the amount of subsequent temper-embrittlement and the rate of embrittlement.

(10) With a constant grain size, neither the austenitizing time nor the quenching temperature affects the subsequent development of temper-brittleness.

(11) The addition of Ti to a Ni–Cr steel did not prevent temper-brittleness from occurring but appeared to increase the temper-embrittlement. A similar effect was found for Al.

(12) A 3% Ni steel was only very slightly susceptible to temper-embrittlement, but a similar steel containing 0·1% of phosphorus was appreciably embrittled by the same treatment.

(13) Temper-brittleness is not the result of a simple precipitation of compounds from solution in the ferrite; it is suggested that it can be explained by the segregation of solute atoms to the grain boundaries in both the austenite and the ferrite.

Acknowledgments

The author wishes to thank Emeritus Professor J. H. Andrew, Professor A. G. Quarrell, and Mr. J. H. Woodhead for their interest and encouragement; Mr. C. Wollerton for carrying out all the analyses; and the Department of Scientific and Industrial Research for a maintenance grant, which enabled him to carry out the research.

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PROCEEDINGS OF THE INSTITUTE

REPORTS OF COUNCIL - OFFICIAL NOTICES

AUTUMN GENERAL MEETING, 1952

THE AUTUMN MEETING OF THE IRON AND STEEL INSTITUTE was held on Wednesday and Thursday, 26th and 27th November, 1952, at the Offices of the Institute, 4 Grosvenor Gardens, London, S.W.1. The President, Captain H. LEIGHTON DAVIES, C.B.E., was in the Chair except during the afternoon of the first day, when his place was taken by Mr. JAMES MITCHELL, C.B.E., President-Elect.

Sessions were held from 10.0 A.M. to 1.15 P.M. and from 2.30 P.M. to 4.40 P.M. on the Wednesday, and from 10.0 A.M. to 12.45 P.M. on the Thursday. A buffet luncheon was provided in the Library of the Institute on the Wednesday.

The Minutes of the previous Meeting, held in London on 30th April and 1st May, 1952, were taken as read and confirmed.

REMARKS BY THE PRESIDENT

The **President**: I am very pleased to welcome, on behalf of the Council, all the members and visitors present. I hope that they will find the papers interesting and the discussions on a high level.

As has already been announced, Her Majesty The Queen has been graciously pleased to become Patron of the Institute. This continues the honour which the Institute has had before. King Edward VII became our Patron soon after his accession, and since then every Sovereign has given us this honour. We wish Her Majesty all good fortune, especially in Coronation Year.

Although I come from Swansea myself, I feel I can say that our meeting there in October was successful. I think that we should place on record our thanks to the Mayor of Swansea, to the Chairman of the Reception Committee, Colonel J. M. Bevan, D.L., M.C., J.P., to the members of the Reception Committee, and to the two Honorary Secretaries, Sir Lewis Jones and Mr. P. O. Williams. (*Applause.*)

FUTURE MEETINGS

The **Secretary** (Mr. K. Headlam-Morley): At the suggestion of the Ministry of Supply, acting in connection with O.E.E.C., a meeting to discuss boron in steel will be held on 18th February of next year. Two American experts are coming over to read papers. Details will be announced later.

The Annual General Meeting will be held at the Offices of the Institute on Thursday 30th April and Friday 1st May, 1953.

Two meetings overseas have been arranged. Jernkontoret invited the Institute to visit Sweden next year, and the Council accepted the invitation with great pleasure, but owing to the Coronation it has been necessary to postpone the meeting. The visit to Sweden is now likely to take place in the second and third weeks of June, 1954. By the kind invitation of Mr. Ingen Housz, of the Royal Dutch Steelworks, it has been possible to arrange a shorter meeting in Holland next year. This meeting will probably be held from 30th September to 6th October, 1953, and will last about 5 or 6 days. The headquarters will be in Amsterdam.

The Engineers Group has continued its very successful meetings. The attendance is usually excellent and discussions free and interesting.

The **President**: We are very grateful to Mr. C. H. T. Williams, Chairman, and to the members of the Engineering Committee for the work that they are doing.

FURTHER REMARKS BY THE PRESIDENT

The **President**: The Council have nominated Mr. James Mitchell, C.B.E., of Messrs. Stewarts and Lloyds Ltd., for election as President at the next Annual General Meeting. I know that this will meet with your hearty approval. (*Applause.*) Mr. Mitchell joined the Institute in 1918 and became a Member of Council in 1944 and Vice-President in 1946. He has served as Honorary Treasurer since 1949. The work that he has done in that capacity has been very valuable. We shall be sorry to lose him as Honorary Treasurer, but he has found someone to take his place. I need hardly say that it will be a Scotsman, Mr. William Barr, who I know will look after our finances very well. Mr. Barr will take over at the beginning of next year.

On the subject of education, I commend to your notice the report entitled " The Education and Training of Metallurgists," prepared by the Joint Committee on Metallurgical Education. It will be well worth your while to read it.

In regard to the membership of the Institute, I want to repeat that I should like to see our membership pass the 5000 mark. I hope you will all help me to achieve that aim, and perhaps go a little beyond it. An appeal has gone out, and if you have not received it please take this as the appeal. In view of the importance of

this Institute and the size of the iron and steel industry, we ought to have a membership of over 5000, and I hope you will all help to see that we get it.

CHANGES ON THE COUNCIL

The **Secretary** reported the following changes on the Council since the last General Meeting in April, 1952:

Vice-President—Mr. C. R. Wheeler, C.B.E.

Member of Council—Mr. B. Chetwynd Talbot.

The following have agreed to serve as Hon. Members of Council during their periods of office as Presidents of affiliated societies:

Mr. F. Kennedy, President of the Cleveland Institution of Engineers, in place of Mr. N. C. Lake.

Dr. C. J. Smithells, President of the Institute of Metals, in place of Professor A. J. Murphy.

Mr. W. D. Jenkins, President of the Ebbw Vale Metallurgical Society, in place of Mr. F. E. Probyn.

Dr. N. J. Petch, President of the Leeds Metallurgical Society, in place of Mr. F. K. Neath.

Mr. W. L. James, President of the Lincolnshire Iron and Steel Institute, in place of Mr. A. Jackson.

Mr. R. S. Brown, President of the Liverpool Metallurgical Society, in place of Mr. V. L. Farthing.

Mr. H. N. Bowen, President of the Swansea and District Metallurgical Society, in place of Mr. L. A. S. Perrett.

Mr. G. H. Latham, a Vice-President of the Institute, has been President of the Newport and District Metallurgical Society since the foundation of the Society and in that capacity has been an Hon. Member of Council. It has now been arranged that the Chairman of the Society, instead of the President, shall be an Hon. Member of Council. The present Chairman is Mr. T. G. Grey-Davies, and he has agreed to serve as an Hon. Member of Council during his period of office as Chairman.

Mr. R. S. Brown and Dr. L. B. Pfeil, representing the Institution of Metallurgists, become Hon. Members of Council in place of Mr. W. E. Bardgett and Mr. W. W. Stevenson.

The following are due to retire at the next Annual General Meeting and are eligible for re-election:

Vice-Presidents—Mr. G. Steel, The Hon. R. G. Lyttelton, Mr. I. F. L. Elliot.

Members of Council—Dr. D. F. Campbell, Mr. G. H. Johnson, Mr. D. A. Oliver, Dr. C. Sykes, F.R.S., Mr. W. Barr.

ANDREW CARNEGIE SCHOLARSHIPS

The **Secretary**: The following award has been made since the last General Meeting:

Mr. A. Shelton (Imperial College). £250 (second grant) for work on the measurement of Poisson's Ratio in the elastic ranges of ferrous–metallic materials and its extension into the elastic–plastic range and fully plastic range at various degrees of work-hardening.

BALLOT FOR THE ELECTION OF MEMBERS AND ASSOCIATE MEMBERS

Mr. A. Jackson and Mr. P. Walker, who were appointed scrutineers of the ballot, reported that the following 169 Members and 30 Associate Members had been elected:

Members

Abdo, Ahmed Fathi (Cairo, Egypt); **Abraham,** David C. (Port Talbot, Glam.); **Addison,** James A., A.M.I. Mech.E. (Glasgow); **Aldridge,** John D. (London); **Andersen,** Otto (Copenhagen, Denmark); **Arkinstall,** Leonard C. (Birmingham); **Arnold,** Kenneth S., B.Sc. (Glasgow); **Ashton,** Murray D. (London); **Aston,** L. W. H. (Billingham, Co. Durham); **Aylmer,** John, A.M.I.E.E. (London); **Baker,** James (Sheffield); **Batey,** Frederic S. (Consett, Co. Durham); **Beach,** Norman, L.I.M. (Chelmsford, Essex); **Beard,** Geoffrey G., A.R.T.C. (Pittsburgh, Pa., U.S.A.); **Benzie,** Gardner W., B.Sc.(Eng.) (Glasgow); **Berry,** Kenneth R. (Sheffield); **Bishop,** Edward, B.Sc. (London); **Blackwell,** Alan J. (Sheffield); **Boustred,** William G., M.A., L.I.M. (Johannesburg, S. Africa); **Bowie,** James B., B.Sc. (Consett, Co. Durham); **Boys,** John R. C. (London); **Bray,** Derek B., A.Met. (Rotherham); **Brunyee,** Horace R. (Sheffield); **Burgess,** Edward Ll., A.Eng. (Scunthorpe, Lincs.); **Burn,** John A. S. (Sheffield); **Burrow,** Harold C. (Bradford); **Carlisle,** Samuel S., M.Sc., A.M.I.E.E. (London); **Carpenter,** Clark B., M.Sc. (Golden, Colorado, U.S.A.); **Chard,** Edward A., B.Sc. (London); **Cheston,** Paul H. (London); **Clarke,** Samuel J., T.D. (Newcastle-on-Tyne); **Cobbett,** Robert F. (London); **Coe,** John J. (Manchester); **Cohen,** Cyril M. (London); **Collins,** Robert D., B.Sc. (London); **Compton,** Joseph N., O.B.E. (Sheffield); **Conard,** George P., Sc.D., M.S. (Bethlehem, Pa., U.S.A.); **Dale,** James R., A.M.I. Mech.E. (Consett, Co. Durham); **Dallas,** Ronald D. F. (Workington, Cumberland); **Davies,** Richard G. (Swansea); **de Jong,** J. J. (Eindhoven, Holland); **Dodge,** Bernard T. (Dagenham, Essex); **Duffey,** James H. (Sheffield); **Dupuis,** Bernard (Boucau, France); **Dutt,** Nihar R., A.I.M. (Burnpur, W. Bengal, India); **Eardley,** Edward S. W. (Birmingham); **Edgar,** Ivan S. (Slough, Bucks.); **Edwards,** Jeffrey R., A.M.I.E.E. (Sheffield); **Ernest,** Walter, A.M.I.Mech.E. (Rotherham); **Ersson,** Bertil (Söderfors, Sweden); **Evans,** Daniel R. (Dagenham, Essex); **Ewan,** George F. (Consett, Co. Durham); **Falla,** Guy, O.B.E. (London); **Fehling,** Hans R., *Dr. Ing.* (London); **Fenn,** Edward M., B.Sc., A.R.I.C. (London); **FitzJohn,** Antony B., A.M.I.Mech.E. (Sheffield); **Fitz-Simons,** Herbert R. (Glasgow); **Franks,** Frederick (Sheffield); **Freebury,** Louis S., A.M.I.Mech.E. (Consett, Co. Durham); **Funder,** Peter (Copenhagen, Denmark); **Gardner,** K. C. (Pittsburgh, Pa., U.S.A.); **Gehm,** Heinz, *Dr. rer. pol.* (Krefeld, Germany); **Golestaneh,** Ahmad Ali, B.Sc. (Manchester); **Goosen,** Johannes J. (Que Que, Southern Rhodesia); **Gray,** Denis (Scunthorpe, Lincs.); **Gray,** William (Bonnybridge, Scotland); **Gruhl,** Wolfgang, *Dr. Ing.* (Aachen, Germany); **Guertler,** William M., *Prof. emer. D. phil. Dr. ing. habil.* (Berlin-Dahlem, Germany); **Gupta,** S. B. (New Delhi, India); **Hale,** William T., B.Sc. (Sheffield); **Harders,** Fritz, D.Eng. (Dortmund, Germany); **Heemeyer,** Wilhelm, *Dipl. Ing.* (Duisburg, Germany); **Henderson,** Arnold (Consett, Co. Durham); **Henderson,** Henry Ira (Consett, Co. Durham); **Hentsch,** Marc (Le Bourget, France); **Heuer,** Russell P., Ph.D., B.S. (Philadelphia, Pa., U.S.A.); **Hewitt,** John L. A.F.R.Ae.S. (Sheffield); **Higgins,** Daniel (London); **Hill,** David G. T. (Bonnybridge, Scotland); **Hitzinger,** Walter, M.Sc. (Linz, Austria); **Hodgson,** William (Dagenham, Essex); **Holman,** Ivo, A.M.C.T., A.I.M. (Birmingham); **Hopkins,** Alan D., M.Sc. (Wolverhampton); **Hutchinson,** Thomas F. (Consett, Co. Durham); **Jameson,** Alfred S. (Chicago, Ill., U.S.A.); **Jeffery,** Derek V., B.Sc. (Johannesburg, S. Africa); **Jensen,** Henry (Bangkok, Thailand); **John,** John H. (Swansea); **Johnston,** Thomas W. (Dagenham, Essex); **Jones,** Lt.-Col. Owain Lewis, O.B.E., T.D. (Birmingham); **Kay,** Sydney, A.M.I.Mech.E. (King's Lynn, Norfolk); **Keeling,** B. S., B.A. (London); **Kempson,** Gerald P., B.A. (Birmingham); **Kerr,** Robert, M.Sc., Ph.D. (London); **Kirkpatrick,** James W., B.S. (Youngstown, Ohio, U.S.A.); **Klepp,** Edwin C., *Dipl. Ing.* (London); **Kozakevitch,** Paul (St. Germain-en-Laye, France); **Lamont,** David H., M.A. (Manchester); **Leach,** John C. C., B.A. (Manchester); **Liander,** Halvard (Västeras, Sweden);

Liptrot, Albert (Sheffield); **Ljunggren,** Stig H. (Kolsva, Sweden); **Lögdö,** John H. (Västeras, Sweden); **Lonnegren,** Karl G., M.I.Mech.E. (London); **Marshall,** Alan (Consett, Co. Durham); **Meachem,** Frederick A., A.I.M. (Wolverhampton); **Mechem,** Dennis V., B.Sc., A.R.S.M. (Birmingham); **Middleton,** John W. (Scunthorpe, Lincs.); **Miller,** Harry N. (Dagenham, Essex); **Miller,** R. (Dagenham, Essex); **Monteil,** J. M. (Paris, France); **Morgan,** George F. J. (Birmingham); **Morgan,** Henry W., M.A. (Swansea); **Nicol,** William A. (Smethwick, Staffs.); **Nightingale,** John, B.Sc., A.M.I.E.E. (Manchester); **Nordström,** Bengt G. O. (Kolsva, Sweden); **Norris,** R. F. (London); **Parish,** Augustus A., B.Eng. (Wollongong, N.S.W., Australia); **Parry,** William D. (Bournemouth); **Paul,** Ronald P. J. (London); **Pearce,** Howard R., B.Sc., A.M.C.T. (Manchester); **Pears,** John (Sheffield); **Pfoutz,** Daniel R., B.A. (Pittsburgh, Pa., U.S.A.); **Pritchard,** William H. (London); **Proctor,** Wilfred E. (Leeds); **Provis,** Anthony (Cardiff); **Rehder,** John E., B.Eng. (Montreal, Canada); **Renouprez,** Jack, A.M.I.Mech.E. (Manchester); **Richards,** Arthur, B.Sc. (London); **Roddy,** Thomas B. (London); **Roux,** Roger (Mondeville, France); **Saklatvala,** B. Sh., B.A. (London); **Samson,** Alexander (Weston-Super-Mare, Somerset); **Schrader,** Hans, *Dr. Ing.* (Jamshedpur, India); **Schwitter,** Charles M., Met.E. (New York, U.S.A.); **Sendzimir,** Tadeusz (Waterbury, Conn., U.S.A.); **Shaw,** Geoffrey, M.A., D.F.C., D.L. (Middlesbrough); **Shaw,** Richard E., B.Sc., F.I.M. (Slough, Bucks.); **Spolders,** Rudolf, *Dipl. Ing.* (Witten, Germany); **Stafford,** Herbert G. (Rotherham); **Stagni,** Otello C., D.Eng. (Milan, Italy); **Stewart,** Gavin B., C.B.E. (London); **Stewartson,** Roy, M.A., A.M.I.Mech.E. (Rotherham); **Stott,** Ronald C., A.I.E.E. (Port Talbot, Glam.); **Stubblefield,** B. M. (East Chicago, Indiana, U.S.A.); **Swindell,** Dennis R. (Port Talbot, Glam.); **Tate,** Norman S. (Consett, Co. Durham); **Taylor,** Hugh G. (Middlesbrough); **Thomas,** Charles E. L. (Cardiff); **Thomas,** Donald R. O., M.A. (Sheffield); **Thwaites,** Colin J., B.Sc.(Eng.), A.R.S.M. (Greenford, Middx.); **Tillberg,** Johan (Trollhättan, Sweden); **Tinley,** Ernest L., M.I. Mech.E. (Wolverhampton); **Towler,** Frank H., M.I. Mech.E. (Leeds); **Towler,** John M., M.I.Mech.E. (Leeds); **Treherne-Thomas,** Robin K. F. C. (Warrington); **Trotter,** David R., B.A. (Scunthorpe, Lincs.); **Tucker,** Walter J. (London); **Underdown,** Harry C. B., B.A., J.P. (London); **Veevers,** J. Foster (Swindon, Wilts.); **Walters,** Llewellyn G. (Port Talbot, Glam.); **Warwick,** William N. (Middlesbrough); **Watkins,** Ernest R., B.Sc., M.I.Mech.E. (Birmingham); **Weitzer,** Helmut, M.Sc., *Dr. mont.* (Linz, Austria); **Wendeborn,** Helmut, *Dr. Ing.* (Frankfurt/Main, Germany); **West,** D. A. (Scunthorpe, Lincs.); **Woodall,** Harold, A.Met. (Sheffield); **Wright,** Wilfred, B.Sc., A.Met., A.I.M. (Sheffield); **Wurm,** Josef (Düsseldorf, Germany).

Associate Members

Argent, Bernard B. (Sheffield); **Bhattacharya,** Kailash Nath, B.Sc.(Met.Eng.) (Kanpur, India); **Chowdri,** Mohammed I., M.Sc. (Wah, Rawalpindi, Pakistan); **Crinnion,** Henry (Middlesbrough); **Crisp,** John (Birmingham); **Cuthbert,** Frank B. (Middlesbrough); **Edwards,** Dennis W., B.Sc., G.I.Mech.E. (Rotherham); **Elliott,** John E. (Darwen, Lancs.); **Evans,** Gerald J. (London); **Faulkner,** Keith E. (Sheffield); **Fenoulhet,** Brian L. (London); **France,** Barry, A.Met. (Sheffield); **Gilbey,** Deryk O. (London); **Griffiths,** David, B.Sc. (Llanelly, S. Wales); **Hacon,** John (Rotherham); **Hardy,** Colin W. (Middlesbrough); **Hewitt,** Jack, A.Met. (Sheffield); **Howard,** William J. (Granville, N.S.W., Australia); **Jee,** Akshoy K. (Orissa, India); **King,** Robert J. (Middlesbrough); **Klein,** Wesley G. (Cowra, N.S.W., Australia); **Macken,** Philip J. (London); **Marshall,** Harold, A.R.T.C. (Scunthorpe, Lincs.); **Meakin,** John (Chesterfield); **Pathy,** R. P. Venkatachala (Banaras, India); **Pickup,** Geoffrey A., B.Met. (Sheffield); **Sharp,** John D. (Cambridge); **Sillitoe,** John F. (Bilston, Staffs.); **Thomas,** Raymond W. (Smethwick, Staffs.); **Wright,** Mark (Liverpool).

The **President** declared the candidates duly elected, and announced that the total membership of the Institute was now 4807.

PRESENTATION OF PAPERS

The following papers were presented and discussed (the dates of publication in the *Journal* are given in parentheses):

Wednesday, 26th November, 1952

Morning Session

Joint Discussion on:

"Cooling of Rimming Steel Ingots in a Casting Pit," by R. T. Fowler and L. H. W. Savage (*July*, 1952).

"Time Studies from Casting to Rolling," by A. V. Brancker, J. Stringer, and L. H. W. Savage (*January*, 1950).

"Mould and Ingot Surface Temperature Measurements," by A. V. Brancker (*July*, 1950).

"Ingot Mould Temperature Measurements," by R. T. Fowler and J. Stringer (*March*, 1951).

Morning and Afternoon Sessions

Joint Discussion on:

"Development of Active Mixer Practice at Appleby-Frodingham," by A. Jackson (*October*, 1952).

"Thermodynamic Properties of Silicon Monoxide," by N. C. Tombs and A. J. E. Welch (*September*, 1952)

"Phosphorus–Oxygen Equilibria in Liquid Iron," by J. B. Bookey, F. D. Richardson, and A. J. E. Welch (*August*, 1952).

"The Free Energies of Formation of Tricalcium and Tetracalcium Phosphates," by J. B. Bookey (*September*, 1952).

"The Free Energy of Formation of Magnesium Phosphate," by J. B. Bookey (*September*, 1952).

"The Effect of Sodium Oxide Additions to Steelmaking Slags. Part II—Dephosphorization of Steel by Soda Slags," by W. R. Maddocks and E. T. Turkdogan (*June*, 1952).

"Phase Equilibrium Investigation of the Na_2O–P_2O_5–SiO_2 Ternary System," by E. T. Turkdogan and W. R. Maddocks (*September*, 1952).

"The Thermodynamics of Substances of Interest in Iron and Steel Making. Part III—Sulphides," by F. D. Richardson and J. H. E. Jeffes (*June*, 1952).

"A Stoichiometric Combustion Method for the Determination of Sulphur in Slags," by C. J. B. Fincham and F. D. Richardson (*September*, 1952).

"Thermodynamic Calculations of Slag Equilibria," by H. Flood and K. Grjotheim (*May*, 1952).

Thursday, 27th November, 1952

Morning Session

"Addition of Boron to Steel by Reduction from Boron Oxide," by G. E. Speight (*June*, 1952).

"Effect of Direction of Rolling, Direction of Straining, and Ageing on the Mechanical Properties of a Mild Steel Plate," by C. F. Tipper (*October*, 1952).

"The Rapid Softening of Cold-Drawn Austenitic Stainless Steels," by F. A. Hodierne and C. E. Homer (*July*, 1952).

DISCUSSION ON PAPERS

DISCUSSION AT MEETINGS · WRITTEN CONTRIBUTIONS

AUTUMN GENERAL MEETING, 1952

THE AUTUMN MEETING OF THE IRON AND STEEL INSTITUTE was held on Wednesday and Thursday, 26th and 27th November, 1952, at the Offices of the Institute, 4 Grosvenor Gardens, London, S.W.1. The President, Captain H. LEIGHTON DAVIES, C.B.E., was in the Chair except during the afternoon of the first day, when his place was taken by Mr. JAMES MITCHELL, C.B.E., President-Elect.

Sessions were held from 10.0 A.M. to 1.15 P.M. and from 2.30 P.M. to 5.0 P.M. on the Wednesday, and from 10.0 A.M. to 1.15 P.M. on the Thursday. A buffet luncheon was provided in the Library of the Institute on the Wednesday.

The Report of the Meeting is published on p. 256.

Discussion on the Paper—

COOLING OF RIMMING STEEL INGOTS IN A CASTING PIT*

By R. T. Fowler and L. H. W. Savage

This paper was discussed in conjunction with the following papers:

"Time Studies from Casting to Rolling," *by A. V. Brancker, J. Stringer, and L. H. W. Savage*†

"Mould and Ingot Surface Temperature Measurements," *by A. V. Brancker*‡

"Ingot Mould Temperature Measurements," *by R. T. Fowler and J. Stringer*§

Mr. **L. H. W. Savage** (British Iron and Steel Research Association) presented the paper, and added: Since this paper was written we have carried out a further series of observations, which have also been published in the *Journal*.[1] This paper shows that a large part of the variation in heat loss can be attributed to the variation in the initial temperature of the mould, which itself can be related to the time since the mould was previously stripped. Figures 10 and 11[1] suggest how this knowledge can be applied as corrections to empirical rules of works practice, thereby reducing factors of safety and fuel requirements.

It will readily be appreciated that we are still a long way from knowing the whole story, and that the numerical values given are relative rather than absolute.

Mr. **N. H. Bacon** (Steel, Peech and Tozer): I propose to confine my remarks largely to the paper by Fowler and Savage. This latest development brings us a little nearer to making practical use of the lessons learned during these investigations. The main result seems to be that non-feeder-head ingots should be charged into the soakers before they have reached full solidification. We know from our own experience that it is quite safe to do this, and it seems to be particularly important since, as Mr. Savage has said, although a delay of 20 min. may mean only a small overall heat loss (about 0·25 therm/ton), the centre temperature falls rapidly and in that

* *J. Iron Steel Inst.*, 1952, vol. 171, July, pp. 277–288.
† *Ibid.*, 1950, vol. 164, pp. 67–84.
‡ *Ibid.*, 1950, vol. 165, pp. 307–313.
§ *Ibid.*, 1951, vol. 167, pp. 283–288.

time can fall below rolling temperature, and considerable time is lost in reheating the ingot.

This finding also confirms the results of Mr. Robinson, of Appleby-Frodingham, who at the Glasgow Meeting showed, from actual measurements at the centre of 10-ton slab ingots, that in 180 min. the centre temperature had fallen to 1200° C. This is in fairly good agreement with the temperatures calculated for the 8-ton ingot and shown in Fig. 11 of the paper by Fowler and Savage.

The authors assume a negligible heat transfer to the top and bottom of the ingot. In fact, however, the heat transferred to the bottom must be of considerable magnitude, because of the chilling effect of the bottom plate. We find this effect of considerable importance, particularly with big forging ingots. Moreover, the heat transfer to the top will be by no means negligible. I feel that further work is urgently required to determine the amount of heat loss.

On p. 286 of their paper the authors state that the 8-ton ingot is completely solid in 160 min., but from Fig. 11 it would seem that complete solidification has taken place approximately 120 min. after filling. Is this discrepancy due to neglecting the heat transfer to top and bottom?

Conclusion (3) states that "The technique using 40 thermocouples in a single mould cannot readily be extended to more than one mould in a cast, . . ." This is very true, "but," the authors continue, "using this correlation a single thermocouple (or contact pyrometer) at the reference point will enable all the ingots to be studied simultaneously, to establish differences between them." Even if such a technique were possible and we could measure the temperature of the individual mould, having obtained that knowledge should we be any better off? I feel that the difference in time in, say, the freezing of an 8-ton ingot within the limits of the mould temperatures that we can use and the steel temperatures at which we can tap and make a good ingot is not nearly so great as the authors' calculations would indicate. The rule that the freezing continues according to the square root of the time is a good working rule, and we have applied it widely. Can the authors please tell us the maximum and minimum times possible, with a reasonable mould temperature, for an 8-ton ingot? To have the mould at 250° C. seems to be asking for trouble. Instead of attempting to find the difference between moulds it would be better to see that they are at a reasonable temperature initially, and therefore I do not think that the single reference point technique will be of much help to us.

So far as the practical application of this work is concerned, it has emphasized the desirability of speeding up stripping and charging. Based partly on my own experience and partly on the knowledge gained from these papers and from the contribution made by Mr. Robinson at the Glasgow Meeting, I have worked out a simple, practical rule which I submit for consideration. If NS is the narrow side of a slab mould, then $NS^2/6{\cdot}7$ will give the desirable time between filling and charging to the soakers; the ingot will then be charged when it is nearly, but not fully, solidified. Thus a 20-in. square ingot should be charged 60 min. after filling, and we know from long experience that this is quite safe. With the 8-ton Summers slab ingot, which is $42\frac{1}{2}$ in. × $25\frac{1}{2}$ in., the formula gives $25\frac{1}{2}^2/6{\cdot}7$, *i.e.*, 98 min., and Fig. 11*b* shows that that ingot is very nearly solid 100 min. after filling. With a 47 in. × 28 in. Appleby-Frodingham 10-ton slab ingot, the formula gives a time of 117 min., and actually their figure is 2 hr.; the data which Mr. Robinson produced at the Glasgow Meeting showed that in 120 min. the centre temperature has fallen to 1390° C.

Mr. **H. Brooke Freeman** (John Summers and Sons Ltd.): On reading the paper by Fowler and Savage, one's first thought is how far the results can be applied quickly to works practice. It is not a question of whether the values obtained are absolute values, but of whether they are sufficiently accurate to reflect the heat changes taking place in a cooling mould and ingot, so that stripping times, in particular, can be related to them. The work has established that the temperature changes at a standardized reference point on the wall of the mould will give a measurable indication of the heat content of a cooling ingot, and this is a big step forward. One can foresee the time when, under standard conditions, a contact pyrometer will be used to indicate the temperature at the reference point, so that stripping time may be decided.

The best time for stripping to take place is when there is sufficient heat left in the still liquid portions of the ingot to equalize the loss of heat that has taken place in the solidified shell and to bring the whole ingot to rolling temperature. If by taking the temperature at a single reference point this optimum time can be regularly and accurately repeated, there will be not only economies in fuel and pit availability but a greater uniformity in the product itself.

In these experiments temperatures have been recorded by inserted thermocouples. I would like to know if there is available a contact pyrometer, sufficiently accurate and robust for works practice, which will measure the mould temperatures not only in the 400–700° C. range but also in the 0–200° C. range, *i.e.*, when the mould is awaiting teeming. The initial temperature of the mould affects the rimming action of the steel very considerably, and the authors suggest on p. 284 that it also affects the heat loss.

The authors have assumed negligible heat transfer perpendicular to the mid-height plane. In a rimming ingot, particularly, there is a vigorous circulation of the liquid metal before the top freezes over, 20–25 min. after teeming, and sectioned ingots suggest that some movement takes place until freezing is complete. The vigorous circulation brings all the liquid metal under the influence of the cooling effect of the bottom plate—weight approx. $4\frac{1}{2}$ tons—and of radiation from the liquid top surface; in addition, considerable heat must be carried away by the CO evolved in the endothermic reaction between C and FeO.

In Table II, trial No. 15, the authors state that the "Test mould held a slagged ingot, and 'blew.'" There was covering of slag on the top of that ingot. In Table V, the heat content of the middle slice was still 231·2 cal./g. 80 min. after teeming, whereas the other ingots in trials 7–14 had reached 232·3–238·1 cal./g. in 60 min. It appears that the slag covering on this ingot has slowed cooling very considerably; if cooling is retarded to the extent indicated by these figures, it is probable that the cooling effects at the top and bottom of the ingot cannot be disregarded in dealing with the mid-height slice.

In Fig. 11 it is indicated that the ingot centre is solid about 100 min. after teeming. Similar ingots tipped on to their sides 60 min. after teeming have not shown in their products either displaced pipe or lamination, which would appear to indicate that solidification was complete 60 min. after teeming.

In Fig. 10 the mean shell thickness is shown as about 4–6 in. after 24 min., and the amount solidified as about 50%. In sectioned ingots we have found a rim zone 5 in. thick after 22 min., and 57% solid. These results indicate, I think, that freezing is somewhat faster than the authors postulate, and this may well be due to the cooling effect at the top and bottom surfaces of the ingot.

Figure 8 shows the temperatures of the inside face of

the mould from the start of teeming. The bank of thermocouples is 36 in. from the bottom of the mould and the ingot is poured to a height of 69 in. The teeming times varied from 2 min. 32 sec. to 3 min. 7 sec., and I am certain that the steel did not reach the level of the couples in 1 min.; the time would be nearer 1½ min., and from Fig. 8 it will be seen that the temperatures are rising steeply at that time.

The temperature of the steel in the moulds taken 1 min. after the finish of teeming is given on p. 285 as about 1530° C., and in Table II temperatures as high as 1552° C. are recorded. I am doubtful of some of these higher temperatures—my own records show temperatures of approximately 1520° C. The lowering of the freezing point, calculated from the percentages of C, Mn, and FeO present and taking into account some segregation, is approximately 15° C., which gives a freezing point of 1520° C. for the steel.

The authors estimate a drop in temperature of 30° C. in the ladle; I would suggest that in future trials they might use dip couples in the ladle and actually measure the temperature.

Table V shows the average heat content of the 9 ingots at tapping to be 316·4 cal./g., with a standard deviation of 1·46, *i.e.*, a variation of approx. ± 1% for 95% of the trials. After 80 min. cooling, the average heat content is 223·5 cal./g. with a standard deviation of 3·72, or approx. ± 3·3%. The average loss of heat between 60 min. and 80 min. cooling is 235·3 − 223·5 = 11·8 cal./g., which, with 95% of the results lying between ± 7·3 cal./g., approximates to a difference of ± 12½ min. in cooling time.

Again, in Fig. 7 at the 500° C. temperature rise of the reference point, the heat lost by the mid-height section is 85 ± 14 cal./g., *i.e.*, a deviation of approx. ± 16½%. In Fig. 11, at 100 min. (*i.e.*, approx. 80 min. after teeming) a variation of ± 14 cal./g. represents a difference of ± 25 min. in cooling time.

It is difficult to assess the above variations accurately from the figures, but they do raise the question of the limits within which the reference point can be used. Table V provides part of the answer; at 80 min. the heat contents of the mid-height slices of seven of the ingots, omitting Trial No. 12 (not recorded) and No. 15 (because the ingot was slagged), varied from 220·1 to 225·5 cal./g., *i.e.*, a variation of ± 2·7 cal./g. This represents a time variation of only ± 5 min. according to the 100-min. line (approx. 80 min. after teeming) of Fig. 11.

If the correlation between the rise in temperature of the reference point and the residual heat content of the ingot is such that stripping can take place at a preselected heat content of the ingot and in a range of time of ± 5 min., then the obvious and important application of these experiments to works practice is at once apparent. Unfortunately, the authors will not be able to test their theories on the same plant, as it is now being dismantled, but I shall look forward with great interest to future work on ingots, especially those cast on bogies.

Sir **Charles Goodeve, F.R.S.** (Director, British Iron and Steel Research Association): There is a good deal of steelworks practice about which I am not at all knowledgeable, but, as I understand the paper by Fowler and Savage, a works will specify a certain time—say 60 min.—before stripping an ingot, and that time will depend on some formula such as that suggested by Mr. Bacon, or on experience, in relation to the sizes of the ingots. This time has been chosen to be well above the critical time before which it is dangerous to lift the ingot by the crane. The authors have translated the time scale into a heat-content scale of the steel ingot and have shown that there is a wide distribution of real values of heat content, and their object has been to reduce the variability. They have a corresponding figure for the heat loss from the molten state, which in one instance they gave as 3½ therms. I am not clear whether the latter corresponded to the standard value or to a critical heat content below which it was not safe to operate.

If one did waste a tenth of a therm unnecessarily, what would this correspond to in the soaking pits? I suggest that it is five times as much; *i.e.*, a loss of 0·1 therm per ton of ingot will need a gain of 0·5 therm in the soaking pit.

Finally, how often, in practice, do accidents occur as a result of being too near the critical time or heat content? And when an accident does happen, how serious is it?

Mr. **A. J. K. Honeyman** (Steel Company of Wales Ltd.): I do not propose to answer the last question, but I would like to make a few practical comments.

The authors refer to the minimum standing time as that which permits the formation of a sufficiently strong skin in the ingot. With bogie casting, particularly with large ingots, the internal condition may be as important. With rimming steels the minimum standing time is that which will ensure such a solid top that there is no danger of 'blowing' during transit. Unfortunately, there appears as yet to be no way of calculating this time with any degree of accuracy, and it is usually based on experience; it may be reduced by quick cooling of the ingot top, by capping, or by water sprays.

Again with bogie casting, in recording track times it is advantageous to divide them into four periods: (i) Standing time, (ii) time taken to arrive at the stripper bay, (iii) arrival at stripper bay to finish of stripping, and (iv) finish of stripping to finish of charging. By this means the source and magnitude of delays can easily be noted, and the effect of standing time on slabbing mill yield and on steel quality can be determined.

In addition to the benefits of closer control of track times discussed by the authors, mention should be made of the effect on ingot moulds. Not only will shorter track times help to increase mould lives but, since the moulds will tend to be cooler when dressed, the dressing will be more efficacious. Moreover, the avoidance of hot moulds will reduce the tendency to form 'risers' when making rimming steels.

When 'bunching' of casts occurs it is essential that preference be given to 'rising' steels, which will require the absolute minimum time in the soaking pits to avoid excessive oxidation of sub-surface blowholes.

Mr. **H. Brooke Freeman**: Sir Charles Goodeve's question is very difficult to answer because, although I have seen thousands of ingots stripped, I do not recall ever seeing one in which the wall—not the top—has burst and molten steel has been thrown out, although several were stripped in less than 1 hr. after the finish of teeming. I have seen 'cobbles' on the ingots, where there has been an exudation near the bottom of the ingot, but this takes place during the first few minutes of teeming. It was because I had seen ingots stripped in less times than are indicated by the authors that I drew attention to all the points which I thought would give an increased rate of heat loss. I think that their times are rather too long, and in actual practice can be cut down. If the ingots are badly wanted by the mill, we strip many casts long before the stated time of 1¼ or 1½ hr., according to which formula is used.

Sir **Charles Goodeve**: If anybody here has seen an accident, will he please come forward?

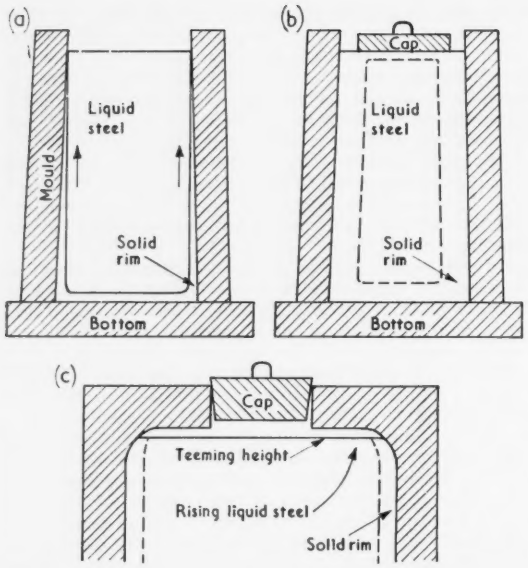

Fig. A—Rimming steel ingots: (*a*) Showing rimming action; (*b*) mould capped to prevent rimming action; (*c*) capped bottle-top mould

Mr. **A. Jackson** (Appleby-Frodingham Steel Co.): I have not seen an accident and would like to explain why they do not occur. If one lifts the mould when stripping and the ingot slides out without it being necessary to press the stripper ram on the top, then it is obvious that the skin is sufficiently strong to enable stripping to take place. With our 10-ton moulds, the average time from tap to strip can be 40 min. and is sometimes less if the heats can be taken from melting shop to stripper sufficiently quickly.[2] If it is necessary to exert pressure on the top of the ingot to strip it, the crust will be broken and molten steel will squirt out in all directions.

A batch of 10-ton ingots stripped in 40 min. must wait at least 2 hr. before they are charged into the soakers, as if they are put in immediately the centre remains liquid so long that 5 or 6 hr. later they will still balloon when put through the mill. To give maximum soaker efficiency the ingots must be cooled for 2 hr. before charging, and then 1½ hr. in the soakers will enable them to be rolled in 3 to 3½ hr. from tapping.

To overcome this difficulty at Appleby-Frodingham, we use a series of gjers pits (unfired), each of which holds four 10-ton ingots. Our practice is to strip the first heat as quickly as possible and charge it into the gjers; the second heat follows and is handled in a similar manner. By the time the third heat is stripped, it is cool enough to charge directly into the soakers and at the same time give the minimum heating time. As this heat is sent to the mill the soakers are charged from the second heat and later the first heat, both of which have been cooling very slowly in the gjers pits.

By this means we can pass an average of 12,000 tons of ingots per week through four rather old-fashioned soakers and with a heat consumption which is always under 1,000,000 B.Th.U.

Mr. **P. Walker** (Cargo Fleet Iron Co., Ltd.): This question of the ingots being liquid takes me back to around 1911, when the late Mr. Benjamin Talbot carried out experiments with the object of reducing piping and dispersing segregate in rail steel. The ingots were stripped 30 min. after teeming, placed in soaking pits for 25 min., taken out and given several passes through the rolls, and then returned to the pits. This operation was carried out while a portion of the centre of the ingot was still liquid. No accident occurred, although some of the ingots did eject liquid from the top end. It may be of interest to mention that this process gave top end rails in which the segregate was present as a thin ring midway between the outside and centre of the section.

More recently, we have had examples of liquid being forced out of the top of low- and medium-carbon ingots during cogging, again without an accident. A feature of this liquid is its high iron oxide content, which is in the region of 80%. This occurrence has been described as soaking pit pipe.

Mr. **R. W. Evans** (Steel Company of Wales Ltd.): I want to try to answer Sir Charles Goodeve's question on the frequency with which trouble is experienced due to metal being ejected from an ingot after it has stood for a short time. I have seen this happen only with a rimming steel ingot, and we believe that the mechanism of ejection is on the following lines. As a mould is filled to the required level, gas is evolved at the solid/liquid interface and passes up the ingot walls (Fig. *A* (*a*)). To stop this rimming action a cap is placed on top, its size depending on the thickness of rim required in the finished ingot (Fig. *A* (*b*)). The object of this is to produce a solid bridge at the top, so that the ingot will be completely sealed off and no further gas will escape; pressure then builds up in the top part of the ingot, and if the pressure becomes too great before the cap is strong enough to resist it there is an ejection of metal which may or may not be dangerous.

This occurs infrequently and in most cases is without danger. In the case of smaller ingots, however, it has happened that, without warning, there has been a shoot of metal as from a pistol; this may travel 30 yards and can be dangerous to a passer-by.

Evidence of the terrific pressures that are built up is shown by our 15-ton bottle-top mould, which is capped very quickly. We teem to the level shown (Fig. *A* (*c*)) and superimpose a thick cast-iron cap, so that little gas is evolved before the ingot is sealed. On occasions the mould has then actually been seen to rise until there has been 6 in. of the ingot showing at the bottom. This is a measure of the pressure that we are trying to suppress by capping a rimming steel ingot.

AUTHORS' REPLY

Mr. **L. H. W. Savage** replied: One of the main points raised by both Mr. Bacon and Mr. Brooke Freeman is heat transfer from the mid-height plane to the top and bottom. Whilst this must occur in an ingot of finite height, we feel that for most of the period with which we are concerned the temperature gradient in a vertical direction near the centre of the ingot will be relatively flat, and, in consequence, heat flow perpendicular to the mid-height slice will be small compared with heat flow in a horizontal direction. Differences between runs on this account should be of a second order and should not affect our main conclusions, which are derived from relative and not absolute values. Our values for heat loss can be taken as minima; they are probably reasonably correct for the first 30 min., but are too low for the later stages.

We propose to continue this work on 15-ton bottle-top moulds, and, in addition to exploration of the mid-height plane, we shall incorporate thermocouples in the top and bottom portions of the mould and in the bottom plate itself, to obtain an estimate of the heat loss.

The surface heat loss from the mid-height section of the mould becomes appreciable after 40–50 min. This we have calculated from the surface temperature, but checking by the temperature gradient in the mould suggests that it may be too low. We are, however, taking all our temperatures on a single recorder graduated from 0° to 1000° C., which means that we are reading the temperatures with an accuracy of about $\pm 2\frac{1}{2}$° C. This is adequate for calculating the heat gained by the mould, but the temperature gradient at the time we are particularly interested in is only of the order of 10° C./in., so that the heat loss calculated from the gradient could be between double and half the mean figure. In our next runs we shall try to obtain a direct measurement of the differential temperature on a separate and more sensitive recorder.

Both Mr. Bacon and Mr. Brooke Freeman have referred to the single reference point. This appears to give a definite indication of the state of the ingot, and we have subsequently used it to compare six moulds in the same cast. Our latest observations suggest that it is reasonably reliable for the first 50 min., *i.e.*, until the surface heat loss becomes of major significance. At present it is not sufficiently accurate to be used completely independently, but we hope that it will become of greater use as the work proceeds.

In regard to Mr. Bacon's point about the time for solidification, the shorter time of 120 min. shown in Fig. 11 occurred with a cold mould, whilst the second figure given, of 160 min., was the average of all 9 runs for which the corresponding average initial mould temperature was 140° C., and was also obtained by extrapolating data from 80 min. In the subsequent paper, to which I have already referred,[1] we have derived an approximate equation relating the solidification time to the initial mould temperature, which gives for this particular practice a range of 120 min. with a cold mould to a little over 160 min. with moulds at 200° C. Because of the assumptions made it appears that both these times are rather too long, but we feel that it is reasonable to expect them to be in the same ratio. In connection with the time of solidification, Mr. Bacon's working rule appears to be very useful and it will be interesting to see how our later work agrees with it.

Mr. Brooke Freeman has raised one interesting point which I had not appreciated myself, that in the last run given in this paper the ingot would have a thin coating of slag on the top which would act as an insulator. I would like to get some further observations, because I feel that in experiments of this nature we cannot place too much reliance on observations of one single run.

The heat of reaction of FeO and C results in a heat loss of about 0·3 cal./g. of steel for each point of carbon removed. We think that, for the quantities likely to be involved, this is less than the probable errors in assumptions about the heat content of the steel at temperatures above 1300° C., and it would be adding undue refinement to incorporate such heats of reaction until we have direct measurement of such heat contents.

In regard to the height of the mould, I was under the impression that we had copied the drawing received from the works, but I thank Mr. Brooke Freeman for the correction.

The time taken for the steel to reach the line of the thermocouples was, I agree, purely an estimate on our part, and we shall stand by the experience of those who are more familiar with the process. The temperature of the steel in the mould is also an interesting point. As Mr. Brooke Freeman says, in one case it was reported as 1550° C.; we were doubtful about this ourselves and we ignored it in saying that in all cases the temperature of the steel was at or about the melting point.

With regard to the temperature drop in the ladle, we have merely made estimations from the formula of Land,[3] although we would have liked to obtain confirmation by direct measurement, preferably by a dip thermocouple somewhere near the bottom in case of possible stratification. Since this paper was written, Hadfield and Donald[4] have published a paper dealing with the direct measurement of steel temperature in 80-ton ladles, which suggests that the average drop in temperature should be of the order of 35° C. instead of our figure of 30° C.

In reply to Sir Charles Goodeve on the question of how much heat has to be used to make up unnecessary loss, this is one of the quantities we want to measure. When cold ingots are heated to rolling temperature in a good soaking pit, the overall efficiency from the time the ingots are charged until they are ready to draw is of the order of 40–50%. In the early stages, however, there is undoubtedly a very high heat-transfer rate and the immediate efficiency may be about 80%, whereas in the last stages of soaking it is probably down to less than 5%. The efficiency for hot charged ingots must be near the tail of this efficiency curve, and I would suggest that, for the condition in which the centre temperature drops below the rolling temperature and any additional heat has to be transmitted back to the centre through the whole thickness of the ingot, about 10 times the actual heat drop will be required (*see* Fig. 11). If the stripping can be done at the right moment, the centre will be losing heat to the intermediate zones at the same time as they are being heated from the outside, but whether there is a sharp break in the heat-content/time curve is a matter that still has to be determined.

To clear up the other point raised by Sir Charles Goodeve, Fig. 3 of our later paper[1] shows a series of lines at 10-min. intervals and a dotted line based on empirical information. We took the figure of 75 min., which gave us 3·5 therms/ton as appropriate for a mould temperature of 220° C., but it might well be appropriate for a mould of 350° C. At the time we made these measurements the works were operating a 6-day week in the melting shop, and the Sunday-night taps were on cold moulds; very often to get the ingots through quickly to the mill they were stripped in under an hour. Generally there is no difficulty, but it sometimes happens that if stripping is started too soon there is not always enough shrinkage for the ingot to come out easily. There may be a tendency, in works which have stripper cranes not fitted with rams, to leave enough time for the ingot to free itself easily.

REFERENCES

1. L. H. W. Savage and R. T. Fowler: *J. Iron Steel Inst.*, 1953, vol. 173, pp. 119–128.
2. A. Jackson: *Ibid.*, 1951, vol. 167, pp. 280–301.
3. T. Land: *Ibid.*, 1941, No. II, pp. 157P–168P.
4. D. Hadfield and A. Donald: *J. West Scotland Iron Steel Inst.*, 1951–52, vol. 59, p. 44.

DISCUSSION ON THE THERMODYNAMICS OF STEELMAKING

This discussion was based on the paper :

" Development of Active Mixer Practice at Appleby-Frodingham " *by A. Jackson,*[1]

which was introduced by the author, and on the following group of papers, for which Dr. F. D. Richardson acted as rapporteur:

" Thermodynamic Properties of Silicon Monoxide," *by N. C. Tombs and A. J. E. Welch*[2]

" Phosphorus–Oxygen Equilibria in Liquid Iron," *by J. B. Bookey, F. D. Richardson, and A. J. E. Welch*[3]

" The Free Energies of Formation of Tricalcium and Tetracalcium Phosphates," *by J. B. Bookey*[4]

" The Free Energy of Formation of Magnesium Phosphate," *by J. B. Bookey*[5]

" The Effect of Sodium Oxide Additions to Steelmaking Slags. Part II—Dephosphorization of Steel by Soda Slags," *by W. R. Maddocks and E. T. Turkdogan*[6]

" Phase Equilibrium Investigation of the Na_2O–P_2O_5–SiO_2 Ternary System," *by E. T. Turkdogan and W. R. Maddocks*[7]

" The Thermodynamics of Substances of Interest in Iron and Steel Making. Part III—Sulphides," *by F. D. Richardson and J. H. E. Jeffes*[8]

" A Stoichiometric Combustion Method for the Determination of Sulphur in Slags," *by C. J. B. Fincham and F. D. Richardson*[9]

" Thermodynamic Calculation of Slag Equilibria," *by H. Flood and K. Grjotheim*[10]

The **Chairman** (Mr. James Mitchell, C.B.E.) said that Mr. Jackson's paper was in the best tradition of the United Steel Companies in providing information about what they were doing, and thanked Dr. Richardson for his admirable presentation of the other papers.

Mr. **W. A. Cameron** (Colvilles Ltd.): I propose to confine my remarks to Mr. Jackson's paper, which, from the details given of modern mixer practice and the record of experimental work on the treatment of molten metal between the blast-furnace and the mixer, is of immediate value to the industry. Some of the experiments on ladle treatment which appeared promising in principle disclosed difficulties and dangers in application, and indicate that many problems remain to be solved before such methods can be regarded as suitable for routine refining in the steel plant.

The development of mixer practice since 1937 is well illustrated in Tables III and IV, and the advantages of modern design and methods are clearly shown. It is interesting to note that the recent results given in this paper indicate that the work done in the mixer is equivalent to a reduction of about 40% in the metallurgical load in the open-hearth furnace. This is based on potential slag volume when using mixer metal as compared with direct blast-furnace iron, assuming that no primary slag is removed. A reduction of this order should mean increased output from steelmaking furnaces, which will vary with the iron burden used, probably being negligible at or below 30% burden but reaching the value of 25% given by Mr. Jackson on a 75% hot metal burden.

Table VI, the effect of mixer slag volume on sulphur loss, gives results which are somewhat unexpected, in that it shows a loss of sulphur from the slag when the slag volume is low and a gain when the volume is high. It seems reasonable to suggest that, since the higher slag volume practice in recent years is made possible by a higher temperature of operation, and since this was achieved by a change in firing practice, the nature of the gas used may provide the explanation. Blast-furnace gas, used during the years when oxide and limestone feeding was low, is, of course, almost sulphur-free, and a sulphur loss from the slag could be expected; but this condition may have changed with the introduction of a proportion of coke-oven gas having an appreciable sulphur content, which may have led to sulphur transfer from gas to mixer slag. That is a simple explanation and it may not fit the facts, but I would like to have Mr. Jackson's comments on it.

Mr. **R. W. Evans** (Steel Company of Wales Ltd.): In reading the papers on thermodynamics, I found myself wondering how we could apply in practice the results which they embody. This problem is probably greater than some of those which the authors had in finding the value of certain constants.

Despite progress in elucidating the physics and chemistry of processes at high temperatures, steelmaking is still largely an art and not a science. A steelmaking operator, each time he taps an open-hearth furnace, has, perhaps subconsciously, to balance a number of conflicting tendencies. He is dealing with four phases: gas, slag and metal, which are molten, and the hearth, which is solid. Equilibrium rarely obtains in these systems, and the composition of each phase may be changing rapidly even while he is deciding his action when he comes near to tapping the furnace; and, to make it more difficult, the four phases are all at temperatures around 1550° C.

The perfection of flow- and temperature-measuring techniques has enabled operators to exercise more control, but even now, 80 years after the first heats were made in Bessemer converters and in open-hearth furnaces, we are still far from being able to make our results scientifically reproducible. Surely, therefore, the steelmaker needs exact qualitative and quantitative knowledge of the high-temperature reactions that govern the efficiency and economy of this process.

The average open-hearth operator has not the time, nor in many cases will he have the technical equipment (by which I mean his own technical equipment), to deal with conceptions such as are embodied in these papers, and therefore we may consider that, after these investigations, another step must be taken, also at a fairly high

[1] *J. Iron Steel Inst.*, 1952, vol. 172, pp. 184–202. [2] *Ibid.*, vol. 172, pp. 69–78. [3] *Ibid.*, vol. 171, pp. 404–412. [4] *Ibid.*, vol. 172, pp. 61–66. [5] *Ibid.*, vol. 172, pp. 66–68. [6] *Ibid.*, vol. 171, pp. 128–136. [7] *Ibid.*, vol. 172, pp. 1–15. [8] *Ibid.*, vol. 171, pp. 165–175. [9] *Ibid.*, vol. 172, pp. 53–55. [10] *Ibid.*, vol. 171, pp. 64–70.

theoretical level, by which these conceptions can be translated into ideas which will definitely help the open-hearth operator.

I was interested in, and often filled with admiration for, the ingenuity which has been shown in a number of cases in the investigations described in these papers.

Referring first to the papers by Tombs and Welch, I assume that it would be possible, in an open-hearth furnace, for the silica roof material to be dissipated as gaseous silicon monoxide. We have to assume that some silicon is present for the reaction $Si + SiO_2 = 2SiO$ to occur. This might, of course, take place by the reduction of silica initially by hydrogen, so that the two stages would be silica + hydrogen giving water vapour + silicon, and silicon + SiO_2 giving SiO. There is some evidence in practice that this does take place. For instance, a silica roof wears rapidly in a reducing atmosphere, and in the early stages of working on a 150-ton furnace fired with 100% cold coke-oven gas the interior surface of the roof rapidly disappeared. It was puzzling to know where it had gone, as the slag composition did not confirm that it had all dropped into the bath. It may have gone off as SiO, which was later reoxidized to silica. If this were possible, it would add greatly to our knowledge of this very worrying feature of silica roof wear.

From the paper on " Phosphorus-Oxygen Equilibrium in Liquid Iron," it seems that phosphorus-bearing iron does not lose its phosphorus to solid lime in the absence of silicon unless the phosphorus content of the melt exceeds that corresponding with the tetraphosphate equilibrium. I do not know whether this result would apply to reactions between two molten phases, but it is certain that (at least in the basic open-hearth furnace) all the silicon, or nearly all, must be oxidized before a start is made on the elimination of phosphorus. Phosphorus is eliminated, therefore, only in the absence of silicon in the metallic phase. Perhaps the authors would comment on this point.

On p. 61 of his paper on " The Free Energies of Formation of Tricalcium and Tetracalcium Phosphates," Dr. Bookey remarks that " To increase our knowledge of phosphorus–oxygen equilibria still further, a study of the free energy of formation of tetracalcium phosphate from lime and from gaseous phosphorus and oxygen was considered to be profitable." The word " profitable " there means, I take it, that it would lead to the next step, which will provide for open-hearth operators some information that they can apply to their process. We operators agree that it is important to know these figures because we hope that, once the free energy has been irrevocably established, it will be used in theoretical studies of the complex slags which we use, as a guide to those slag-making materials that we should use and when to use them. The free energy of formation of tetracalcium phosphate by itself is no more use to the open-hearth manager than the nitrogen in his combustion air; it will be the further steps, making use of these important constants, which will shed light on practical applications, and we therefore look forward to developments in this work.

In Dr. Bookey's second paper, on " The Free Energy of Formation of Magnesium Phosphate," the practical problem is stated with greater clarity. Despite numerous investigations of the equilibria between liquid iron containing phosphorus and slag of various complex constitutions, the relative rôles of lime and magnesia in effecting dephosphorization are still not clear. It will be an advantage to elucidate this question. Indeed, we may find the answer by a practical experiment in the open-hearth furnace. We would deliberately make a high-magnesia slag, allowing the MgO to replace CaO, and note whether phosphorus was in fact more easily removed, always assuming that we could maintain the slag in a liquid condition. In the manufacture of low-carbon steels, the oxygen content of the system, both slag and metal, is high enough to eliminate the problem of dephosphorization with normal slag. With high-carbon steels, however, the problem is not so much to reduce the phosphorus in the bath before tapping as to prevent with certainty its reversion back to the metal during tapping, and in these problems temperature plays a very large part.

As a practical man, I view the two papers on soda-slags with certain misgivings, bearing in mind the bad effect of soda on nearly all steelworks refractories. Is this pure or applied research, and, if the latter, to what sphere of steelmaking is it thought that the results might apply ? On p. 136 the authors state " on the other hand it is neither economical nor practicable to introduce other bases in place of lime, *e.g.*, oxides of iron, manganese, or magnesium. The slags that brought about the best phosphorus removal, down to 0·02–0·01% of phosphorus, contained large percentages of iron (*see* Table II) and cannot be considered suitable for practical application." In normal open-hearth steel manufacture we reduce phosphorus to 0·01–0·02% with slags containing 20% and more of total Fe.

The paper on the thermodynamics of sulphides, by Richardson and Jeffes, again provides knowledge which will be useful if further steps can be taken to elucidate the mechanism of desulphurization in the steelmaking processes. As steelmaking operators, we cannot at this stage take advantage of the knowledge that is provided in their paper.

Mr. **D. A. Oliver** (B.S.A. Group Research Centre): Sooner or later the practical steelmaker will have to become accustomed to thermodynamic notations. A few years ago we obtained an English translation of Schenck, which had the great advantage that when one made a substitution in a formula that was not really accurate, theory could be correlated with practice to a considerable degree. New researches have shown where Schenck was wrong, and I regard the papers now being presented as important parts in a new story; but the problem still remains as to the extent to which these formulae can be applied in practice. There is also the question of notation. It is a tragedy that thermodynamics has managed, somehow or other, to give birth to about three kinds of algebra, and the moment one tries to apply them one finds that the authors are using unfamiliar symbols, which tends to hold up the much-needed advance in practical application.

Referring to the paper by Tombs and Welch, it is stated in their General Conclusions (p. 78) that: " The results now reported establish that silicon monoxide may be expected as a normal product of reduction of silica (or silicates) at temperatures in and above the steelmaking range, and the thermodynamic data given are sufficient to predict the behaviour of silicon monoxide in at least the simpler reactions." Unfortunately, the steelmaker does not encounter simple reactions, and in view of the increased use of the oxygen lance causing higher temperatures, do they agree that we may expect to get far more silicon monoxide produced in steelmaking in the future ?

The interest of this reaction appears to be fundamental to a modern discussion of the acid open-hearth process, or of the acid electric process. I would like to ask how the authors regard the application of their data to the simplest possible examination of the acid open-hearth process. Normally, the chemical analysis of an acid open-hearth slag is returned as weight per cent. of the

Table A

EFFECT OF MANGANESE, AND CONDITIONS ARISING DURING ACTIVE MIXER REPAIRS

Period	Mn in Pig Iron, %	Charge to Mixer				Molten Iron, Mixer to Furnaces			At Slagging		Finished Steel	
		Composition, %	Si, %	S, %	Mn, %	Si, %	S, %	Mn, %	No. of Charges	No. at 0·065% S or Less	No. of Casts	No. at 0·045% S or Less
1	1·13	Molten iron 87·02 Cold pig iron 2·98 Scrap 10·00	0·93	0·093	1·06	0·45	0·055	0·80	194	149 (= 76·8%)	554	381 (= 68·8%)
2	0·67	Molten iron 93·51 Cold pig iron 0·25 Scrap 6·24	1·11	0·100	0·670	0·54	0·072	0·52	164	57 (= 34·7%)	552	171 (= 30·3%)
3	0·83	Mixer off for repairs				1·04	0·109	0·83	186	17 (= 9·1%)	505	36 (= 7·13%)

oxides FeO, MnO, CaO, and SiO_2. According to Schenck, if there is 15% of FeO, about 10% (say) of it is tied up with some of the other oxides and about 5% can be regarded as free. This implies that the oxygen in the steel is $L \times 5\%$ (free) FeO, where L is the partition coefficient between the slag and the steel. In practice we never know, even if we have the correct total FeO, exactly how the partition occurs between the 'dead' FeO and the 'active' FeO. Perhaps Dr. Richardson or the authors would suggest how new data should be used for this purpose, bearing in mind the presence of SiO as well as SiO_2. Perhaps then we may be able to define what further work is necessary to enable a complete analysis of the acid process to be put on a sound basis.

Mr. **P. Walker** (Cargo Fleet Iron Co., Ltd.): In looking through Mr. Jackson's paper I expected to find something which might apply to our practice. The two items on which I concentrated were the effect of the manganese content of the blast-furnace iron delivered to the active mixer, and the conditions which arise in a hot-metal open-hearth shop when the active mixer is off for repairs.

Table *A* illustrates the effect of these two factors on the sulphur content of our steel furnace charges and the finished steel. In the first two periods the active mixer was in operation, and in the third it was off for repairs. In the latter period, the molten blast-furnace iron to the steel furnaces shows pronounced increases in both silicon and sulphur contents.

The description 'At Slagging' refers to the period in our hot-metal tilting practice during which we take slag off and put on new lime to form the finishing slag. It is our experience that, if the sulphur content at this stage is 0·065% or less, we have no trouble from sulphur during the final refining period.

These figures show that the higher manganese iron (1·13% Mn) used in Period 1 has resulted in a definite improvement in regard to sulphur over that for Period 2, when the iron contained 0·67% Mn. In Period 3 the metallurgical load on the steel furnaces was appreciably increased as a result of the much higher silicon and sulphur contents of the iron used.

Apart from the benefit of less trouble from sulphur during working of the charges in Period 1, an improvement in the surface of the rolled sections occurred due to the lower sulphur content of the finished steel, which in a number of casts was under 0·035%.

In regard to the manganese content of blast-furnace iron, there appears to be a tendency to reduce the percentage of this constituent, probably because of the much increased cost of manganese ore. For example, in the Symposium on Steelmaking in 1938,[1] seven out of eight producers used iron containing 1·20–1·40% Mn.

I would like to ask Mr. Jackson the minimum amount of manganese which he would consider to be suitable for his process, and what modifications, if any, he makes in his practice, either at the blast-furnaces or in the melting shop, when his active mixer is off for repairs.

Dr. **V. Giedroyc** (Blast-Furnace Research Department, Imperial College of Science and Technology): The rôle of the active mixer in lowering the metallurgical load on open-hearth furnaces is mainly limited to the removal of silicon and manganese. The amounts of phosphorus and sulphur removed are generally small. This limitation with respect to sulphur, which is mainly due to oxidizing conditions prevailing in the mixer, cannot be changed without a profound negative effect on the removal of silicon. It is true that flotation of manganese sulphide can lower sulphur in the metal, but this method is expensive and wasteful. It becomes increasingly important, therefore, to remove sulphur before the metal is transferred into the mixer. Many steel plants have already been doing this by using the soda-ash process, with varying degrees of success. An alternative method of desulphurization, using solid lime, would seem to be more satisfactory, but it has not yet been sufficiently tried on a works scale. Not only can the problem of sulphur removal be solved by the use of solid lime, but also the problem of removing the last traces of blast-furnace slag. In Mr. Jackson's paper, the harmful effect of blast-furnace slag finding its way into transfer ladles and further into the mixer is particularly stressed.

If a teapot-type ladle filled with lime-coated coke is placed between the transfer ladle and the blast-furnace runner (*see* Fig. *A* (*a*)), not only would considerable

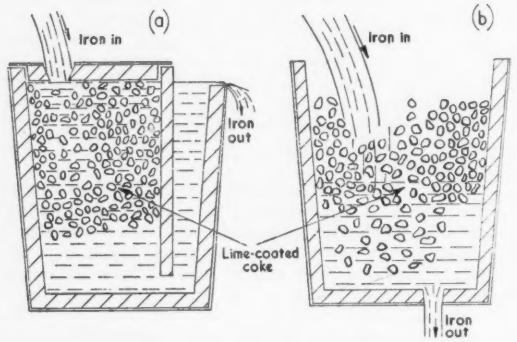

Fig. A—Desulphurizing ladles: (*a*) Teapot-type ladle; (*b*) bottom-pouring ladle with open nozzle

desulphurization take place but the blast-furnace slag would be retained in the filter. However, a less elaborate method, using an ordinary bottom-pouring ladle with an open nozzle and containing lime-coated coke, could be employed (*see* Fig. *A* (*b*)). By pouring the iron through such a vessel, similar purification can be obtained. Extensive laboratory experiments[2] followed by some small-scale works trials have shown that this method is promising, and it is hoped that further large-scale trials will be carried out.

The paper by Maddocks and Turkdogan on the effect of sodium oxide additions to steelmaking slags, particularly Part 1,[3] is of importance to those who use soda-ash for desulphurization in ladles. On the basis of their results the authors showed that basic slags (containing calcium) cannot retain sodium oxide. The mechanism of the reaction is not certain but, whatever it is, the graph presented by Maddocks and Turkdogan (*loc. cit.*,[3] p. 257, Fig. 8) should be useful to those who use the soda-ash process. In this graph the sodium-oxide content of the slag was plotted against the lime/silica ratio in the slag. When this ratio was above 0·24, the sodium oxide content of the slag was negligible.

Although the soda-ash process is widely used, the practice of its application is often the source of more trouble than the unavoidable minimum. In works where soda-treated iron is transferred to the open hearth without careful skimming, severe roof corrosion occurs. This is not observed when the pig iron enters the open hearth *via* a mixer. Although, in both cases, the residual soda-slag mixed with a limey slag causes the evolution of sodium, the lower temperature in the mixer moderates the corrosive effect.

In the light of the work of Maddocks and Turkdogan, the practice of some firms of using soda-ash mixed with lime seems to be correct, because when a limey slag is formed the unreacted sodium oxide cannot be retained and is at least partially volatilized from the ladle. In this way some soda is removed from the system and therefore refractory attack is diminished. In view of these facts it is clear that further study is necessary to determine the optimum conditions for the soda-ash process.

The **Chairman**: I feel that Mr. Jackson's paper and the group introduced by Dr. Richardson are really two different aspects of the same subject. The theoretical papers postulate a state of equilibrium, and the authors perhaps have overlooked the great difficulty of bringing this about, *i.e.*, of getting a sufficient intimacy of mixture between the reacting bodies. Probably the classic instance of this difficulty occurred in the Perrin Process. Some of the apparent confusion of thought may be due to the fact that this difficulty is frequently overlooked in theoretical discussion.

Dr. **J. Pearson** (British Iron and Steel Research Association): In regard to the chemical composition of open-hearth slags, Mr. Oliver pointed out that Schenck regarded some part of the iron oxide as bound and the remainder as free, and that, by using the oxygen partition coefficient, the oxygen content of the metal could be calculated from the concentration of the free ferrous oxide. In the example quoted by Mr. Oliver there is no need for concern as to the possible influence of silicon monoxide; a slag with 15% of FeO would be too oxidizing to permit the existence of SiO.

It has now become customary to express the proportion of a constituent in the 'free' state by use of the term 'activity,' so that, in Mr. Oliver's example, one would refer to an activity of 0·33 for FeO. Dr. Turkdogan and I have recently studied the effect of temperature and slag composition on the activity of FeO, and the results will be published shortly. We have treated steelmaking slags as belonging to a pseudo-ternary (CaO + MgO + MnO)–FeO–(SiO_2 + P_2O_5) system, because it appears that, so far as their influence on the activity of FeO is concerned, MgO and MnO are equivalent to CaO. On the ternary diagram have been plotted FeO iso-activity curves from which it is possible to read the activity of FeO for any slag composition. The use of this value with the oxygen partition coefficient for the prevailing temperature will then give the equilibrium oxygen content of the metal.

From Dr. Bookey's results we can make the following calculation:

$$P_2 + 2\tfrac{1}{2}O_2 + 3CaO = Ca_3P_2O_8$$
$$\Delta G_T^\circ = -537\cdot0 + 0\cdot13T \text{ kg.cal.} \quad ...(1)$$

$$Mg_3P_2O_8 = 3MgO + P_2 + 2\tfrac{1}{2}O_2$$
$$\Delta G_T^\circ = +483\cdot0 - 0\cdot13T \text{ kg.cal.} \quad ...(2)$$

$$Mg_3P_2O_8 + 3CaO = Ca_3P_2O_8 + 3MgO$$
$$\Delta G_T = -54\cdot0 \text{ kg.cal.} \quad .....(3)$$

The equilibrium constant K_3 of reaction (3) thus has a value at 1600° C. of antilog 54,000/(4·575 × 1873) = 2 × 10^6; *i.e.*, the ratio of magnesium phosphate to calcium phosphate is approximately one-millionth of the ratio of magnesia to lime in the slag. In principle, therefore, magnesia will react hardly at all in a lime-bearing slag. This conclusion is derived, of course, from results obtained in a study of the reactions of solid phosphates, and may require some modification when slag melts are concerned, but it is in line with the findings of Flood and Grjotheim. The equation (*see* p. 69 of their paper)

$$\log K'^{(2)} = 21N_{Ca^{++}} + 12N_{Fe^{++}} + 18N_{Mg^{++}} + 13N_{Mn^{++}} \quad .....(4)$$

indicates that in molten slags the value of K_3 is approximately 10^4. This is smaller than the figure calculated above, but it still indicates that magnesia can play only a small part in dephosphorization in a lime-bearing slag. The values of 21, 12, 18, and 13 in equation (4) may be regarded as measures of the relative stabilities of the phosphates of calcium, iron, magnesium, and manganese, and it is thus clear that, if magnesia plays little part in dephosphorization, manganese oxide and ferrous oxide are of even less importance. For dephosphorization, therefore, the operative needs to know only the lime content or activity of the slag and can regard magnesia, manganese oxide, and iron oxide as inert. Because the ferrous oxide in the slag may control the oxygen content of the metal, this oxide may be important in dephosphorization, but only in so far as it helps to produce P_2O_5.

Dr. Richardson related the action of phosphorus as a deoxidant to the possibility of removing phosphorus in the presence of carbon. He implied that if an attempt were made to remove phosphorus by oxidation the oxygen would combine preferably with the carbon, and that the carbon would have to be reduced to a low level before the phosphorus could be removed. This is no doubt true if the whole system is in equilibrium during the process, but, as Mr. Mitchell pointed out, the systems are not necessarily in equilibrium and the reverse could happen, *i.e.*, the carbon could be removed first. Certain operatives at Linz and Donawitz, in Austria, claim that by blowing high-pressure oxygen on to iron in their converters they can reduce the phosphorus very quickly, wait for the carbon to come to the desired level, and then tap. In this instance, therefore, the factor deciding whether carbon or phosphorus will be removed first is a kinetic one, in the sense that a reaction which appears

thermodynamically most probable does not occur first because of some interfering factor, and a reaction which is thermodynamically less likely proceeds because the existing conditions are in its favour.

The **Chairman**: Dr. Pearson has raised a very interesting point in commenting on Dr. Richardson's remarks about the selective oxidation of carbon and phosphorus. The classic commercial instance is the Bessemer converter.

Dr. **A. H. Leckie** (British Iron and Steel Research Association): I was very glad to hear Mr. Mitchell deprecate the tendency to place Mr. Jackson's paper in a different category. The connection between Mr. Jackson's paper and the others is really very much closer than most people would gather from the discussion. Some of his results, quoted in his list of mixer analyses, should be studied very carefully by 'fundamental' chemists, because there are some very useful data to be found in them.

Mr. Jackson, who is a strong protagonist of the active mixer, mentioned his experiments with ladle desiliconizing. He tried the use of scale in the ladle and had trouble with skulling, also with proper slag formation. He also used oxygen in the ladle, but apparently he has not used both oxygen and scale at the same time. I would like to call attention to the experimental work that is being carried out in at least one plant (and probably others) in this country, where scale, with lime and fluorspar, and oxygen are being used in the ladle together, with considerable success.

Mr. Jackson's paper also emphasizes the need for the development of some method of separating slag and hot metal in a cylindrical vessel. The success of many of the alternative pre-refining processes must depend on this, and chemical engineers should attack the problem.

With regard to Mr. Jackson's figures for dephosphorization obtained in the mixer, certain of the analyses show reductions in phosphorus from 1·5% to 1·2%. Presumably the carbon content is over 3%. I think, therefore, that the question of what dephosphorization can occur in the presence of carbon should be studied in the light of these analyses of metal and slag—not, perhaps, of those given in the paper, which seem to be averages, but of figures for some of the individual taps, where the degree of dephosphorization may be greater than is shown by the average figures.

Maddocks and Turkdogan refer to what they call the dephosphorization of iron, but since the carbon content is so low, it can hardly be called iron; it is almost a high-carbon steel. The strong desulphurization effect observed is also noteworthy, as it shows the advantage of soda-ash for desulphurizing, and suggests the use of a mixture of lime and soda-ash.

Dr. Bookey's thermodynamic figures for tricalcium and tetracalcium phosphates suggest that tetracalcium phosphate is more stable than tricalcium phosphate. According to Herasymenko and Speight,[4] however, tricalcium phosphate gives a tetrahedral phosphate ion which one would expect to be a relatively stable unit compared with the phosphate component of tetracalcium phosphate, as the latter is difficult to visualize in terms of ionic structure. Perhaps Dr. Richardson would comment on this.

Richardson and Jeffes' Fig. 2 is very important from the point of view of desulphurization. Cerium is obviously the most potent desulphurizer, but some other metals, such as calcium and magnesium, are also potent desulphurizers, as can be seen from Fig. 2; sodium probably is not, by the time that these high temperatures are attained. Magnesium has been tried with considerable success, particularly by the Department of Mines in Canada.[5] It will desulphurize iron very effectively, the violence of the reaction being controlled by additions of lime, but apparently it is too costly, even when the amount of magnesium required is lessened by the use of lime. I believe this is also the main objection to calcium. Both calcium and magnesium have boiling points rather close to the temperature of the pig iron, so that they vaporize readily. Cerium has a higher boiling point, which explains why it is possible to get it into the iron rather more readily, as shown by Mr. Morrogh in his nodular iron experiments.

Aluminium may have possibilities; however, it is shown on Fig. 2 merely as a dot, and I would like to ask Dr. Richardson whether the entropy term is sufficiently well known to expand this into a line. If the line is drawn from information given in the paper, it seems to rise rather steeply. It may be so steep that it crosses the iron sulphide line at a low temperature, so that it may not be possible to use aluminium for desulphurizing. I do not know what happens when aluminium is put into pig iron, but it might be worth trying. Perhaps an objection is that aluminium sulphide might not be separated easily from the metal.

If the iron sulphide line at the top of Fig. 2 is extended slightly, it will cross the $2C + S_2 = 2CS$ line at about 2000° C., which suggests that if the iron could be heated to a sufficiently high temperature it would desulphurize itself by virtue of its contained carbon. The 'blast-furnace reaction' would take place but with sulphur instead of oxygen.

Dr. **T. P. Colclough** (British Iron and Steel Federation): It is of special importance to note that the fundamental work which has been undertaken in B.I.S.R.A. during the last few years has now reached the stage at which application to everyday practice is possible.

The paper by Mr. Jackson has emphasized that it is, apparently, impossible in an active mixer to desiliconize and desulphurize the metal at the same time. In view of the increasing amount of pig iron that must be used per ton of steel made, it is of increasing importance to focus attention on methods, probably separately, of effecting both a reduction of silicon and a reduction of sulphur.

With regard to the relative rates of oxidation of phosphorus and carbon, I consider that there is a lack of appreciation of the factors involved. The rate at which phosphorus is oxidized in the bath is determined by the amount of lime that is available, as well as the available oxygen. The work of Dr. Pearson and his colleagues has established the ratio in which CaO, P_2O_5, and SiO_2 combine, and also the relative stability of the phosphates of lime, magnesia, and manganese. This work has confirmed the views established by empirical practice, that the effective ratios were $2CaO.SiO_2$ and $3CaO.P_2O_5$. How were these established? In 1923[6] I described how lime and iron oxide were fed to the molten bath of steel and the reactions allowed to proceed to completion. At this stage the above ratios of CaO, SiO_2, and P_2O_5 were established in the slag; then silica was added to the slag and immediately phosphorus returned from the slag to the metal until the previous balance of $2CaO.SiO_2$, $3CaO.P_2O_5$ was re-established. This reversal of phosphorus oxidation leads one to disagree entirely with the diagram in the papers submitted, with regard to the relative oxidations of carbon and phosphorus. The relative rates of removal of these two elements in a steel bath are governed by the basicity of the slag.

In a paper to the Faraday Society in 1925,[7] I submitted evidence that with a slag of low basicity practically the whole of the carbon is oxidized with a very slow elimina-

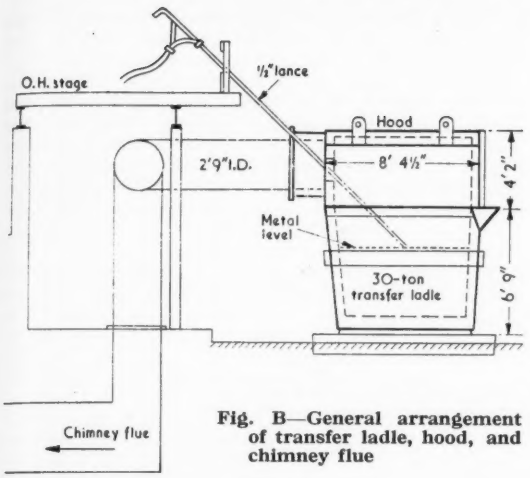

Fig. B—General arrangement of transfer ladle, hood, and chimney flue

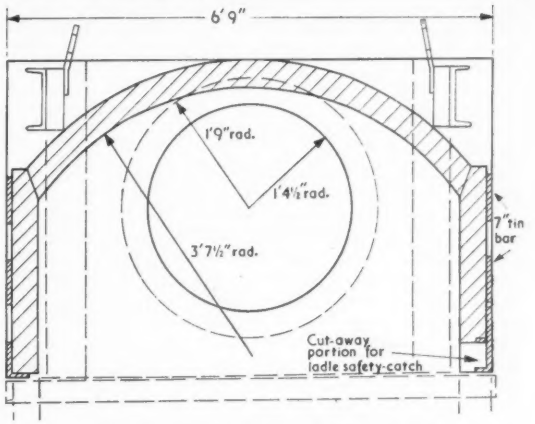

Fig. C—Construction of fume hood

tion of phosphorus, but with a slag of high basicity, *i.e.*, with free lime, practically the whole of the phosphorus is removed while the carbon remains at 1%. These experiments clearly demonstrated that, provided there is sufficient lime to combine with the P_2O_5 which is formed, the phosphorus in the metal will be oxidized in preference to the carbon over the complete range of carbon content.

I suggest that Dr. Pearson and his colleagues might investigate this question further, with particular reference to the basicity of the slag.

Mr. **D. J. O. Brandt** (British Iron and Steel Research Association): I would like to mention what we are doing, on a small scale, on the vexed question of desiliconization. Mr. Jackson showed that it was not easy to desiliconize with oxygen or iron oxide scale in his ladle or in his mixer, owing to mechanical difficulties, and he said that if he tried lancing oxygen into the ladle there was considerable difficulty with fumes. Up to a point these difficulties have been overcome on a small scale at the works of the Brymbo Steel Company, where the shop has to be operated on direct metal because there is no mixer. The small blast-furnace taps only 18–20 tons at each tap and makes only about 100 tons per day, and the open-hearth furnaces are between 40 and 80 tons capacity. Each furnace takes about 20 tons of hot metal, and the remaining pig charge (about 50%) is added cold. Over the past year Mr. Emrys Davies has successfully developed there a system of desiliconization in a special transfer ladle, using an oxygen lance, before the metal goes into the open-hearth furnace. The set-up consists simply of this ladle (*see* Fig. *B*), which is one-third larger than is necessary and which, when containing 20 tons, is filled up to the dotted line, and the use of an oxygen lance. To overcome the difficulty of fumes, a hood fits over the ladle; it is not fixed in any way, but stays there by the force of gravity. A hole in it admits the oxygen lance, and on one side there is a vent which is lined with ganister fireclay (*see* Fig. *C*). The ladle is set down in a special place on the shop floor, using a locating stud, and the vent in the hood matches up with a vent leading to an old open-hearth stack, no longer used as such. The draught created by the stack is sufficient to draw off the fumes.

By means of the oxygen lance, inserted as shown in Fig. *B*, oxygen is pumped into the metal. The silicon, initially 1·0–1·3%, is reduced to about 0·6%. About one-quarter to one-third of the manganese (initially about 1·3–1·5%) is lost, and at the same time a significant amount of sulphur can be eliminated, chiefly because before the metal is charged to this reaction ladle some limestone is put in the bottom, with a little fluorspar to help it to flux down.

When we started this process we used to add soda-ash also, to assist slag formation, but when we stopped adding soda-ash we did not get any less desulphurization. Without the use of soda-ash the sulphur can still be removed from about 0·05% down to 0·03% regularly.

This method is very convenient, as a considerable quantity of scrap can be melted in this ladle, including ladle skulls, which would otherwise have to be taken to the end of the shop and cut up with an oxygen torch before being put into the charging boxes. In addition, the output of the open-hearth furnace has increased and the fettling charges are reduced. The metal temperature is increased by 200° C., giving quick slag formation in the open hearth and a shorter waiting period before refining can begin.

The important point, however, is that this has been done only on a small scale, and it is doubtful whether the difficulties which would be encountered on a large scale (in trying to desiliconize, say, 80 or 100 tons) would be easily overcome. When oxygen is blown into a metal containing a high quantity (over 1%) of silicon, 'slopping' occurs; this might be a serious difficulty in a large ladle with an open top, just as it can be a serious nuisance in the Bessemer converter, but we should not overlook the fact that the degree of mechanical mixing which is obtained in this way is of considerable help, as it assists the oxygen to work more efficiently.

We in B.I.S.R.A. feel that in practice it might be advisable, when undertaking work on a bigger scale, to

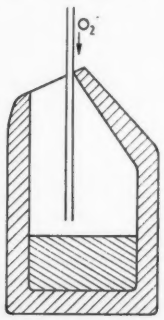

Fig. D—Method of oxygen-blowing in converter at Linz

tap off a fifth or a quarter of each blast-furnace cast and blow this aliquot portion with oxygen as described; then, having reduced the silicon more than is necessary and raised the oxygen content, transfer this metal into another ladle (or into the mixer) and allow the oxygen it now contains to continue the reaction. We are still investigating this problem.

Dr. Pearson, if I understand him correctly, said that at Linz, where certain operators use a converter which is blown with oxygen from above in the method shown (*see* Fig. *D*), the fact that they use oxygen enables them to reduce the phosphorus before the carbon, because it alters the reaction 'kinetics.' I suggest that the fact that oxygen is used in the conversion is really irrelevant; all that they have done is to make the basic converter work in the same way as the basic open-hearth furnace. In the latter, the melters take care that the slag is sufficiently limey and properly melted, so that they can be certain of getting the phosphorus away before they go on to the carbon boil. In the ordinary converter, air is blown in from the bottom, and the phosphorus cannot be reduced until the carbon has gone, because the slag is not formed until carbon has been eliminated. After the 'foreblow' in the basic converter (*i.e.*, the silicon and carbon removal period), lumps of unmelted lime are still floating on top of the slag, and only when carbon is out can oxygen reach the top of the metal and the lime be fluxed down, whereupon calcic ferrite is formed and the lime enters the reacting phase. Then, and only then, can the slag hold the phosphorus as calcium phosphate.

In the converter at Linz, iron oxide is formed immediately the oxygen jet strikes the surface of the metal; this immediately fluxes the lime, so that the slag is formed directly, and (as in the basic open-hearth furnace) phosphorus will reduce at the same time as the carbon. This point has been ingeniously demonstrated recently at Oberhausen, in Germany, and was described by Bulle.[8] A small 1-ton experimental converter with a water-cooled bottom has been constructed and this is (bottom) blown with 100% pure oxygen. Now if this vessel is operated in the normal way with burnt lime to form the slag, it will refine iron in the same way as an ordinary basic converter, and the phosphorus will not reduce until the carbon has gone; but if artificially melted calcic ferrite is added during the operation of the converter—*i.e.*, deliberately making the slag outside and adding it during the blow—then the phosphorus will go out with the carbon. I do not think the fact that *oxygen* is used has anything to do with the way in which the phosphorus and carbon react in that converter. The vital point at Linz is that the *slag* is formed early in the course of the blow.

The **Chairman**: We found exactly the same thing about fifteen years ago. There is no real formation of the Bessemer slag until the carbon has gone.

Dr. **A. W. Richards** (Imperial Smelting Corporation Ltd.): Richardson and Jeffes, in their paper on sulphides of interest in iron and steelmaking, are to be congratulated on putting together this information in such a way that chemists and metallurgists can have easy access to the data. In their notes on zinc sulphide (p. 174) they refer to the calorimetric work of Kapustinskii and Korshunov,[9] who measured the heat of formation of zinc sulphide and obtained the following results:

$$\text{Zn (solid)} + \text{S (rhombic)} = \text{ZnS (solid)}$$

$\Delta H_{298} = -45{,}200$ cal./mole (wurtzite)
$\Delta H_{298} = -48{,}400$ cal./mole (sphalerite).

We shall be concerned here with the result for sphalerite, although the large difference in the two heats of formation is almost certainly anomalous when one considers the small change in structure of the two forms of zinc sulphide and their similarity in compressibility.

Using the above heat of formation and known thermodynamic data, Lumsden[10] has calculated the expected dissociation pressure of zinc sulphide. We have since carried out work at Avonmouth on the vapour pressure of zinc sulphide by entrainment of the vapour in an inert gas. When sulphur is added to the inert gas passing through the zinc sulphide, the zinc pressure is depressed to an extent indicating complete dissociation of the vapour at 1000° C.

The direct measurement of zinc pressures by this entrainment method leads to a value of $\Delta H_{298} = -48{,}000$ cal./mole for the heat of formation of sphalerite. This has been confirmed to within 1 kg.cal. by measuring the equilibrium in the reduction of zinc sulphide with hydrogen and utilizing the known thermodynamic data for hydrogen sulphide.

The work of Veselovskii,[11] who measured zinc-sulphide vapour pressures by an effusion method, can be interpreted from the point of view that he was actually measuring a dissociation pressure, *i.e.*, the pressures of zinc and sulphur and not the pressure of zinc sulphide. On making this correction to his results we arrive at a figure of $\Delta H_{298} = -49{,}300$ cal./mole for the heat of formation of sphalerite.

There is no mention of tin sulphide in the collection of data on sulphides by Richardson and Jeffes, and I would therefore like to quote a figure for the free energy of formation of tin sulphide which we have obtained from work on the reduction equilibrium of tin sulphide with hydrogen:

$$2\text{Sn (solid)} + \text{S}_2\text{ (gas)} = 2\text{SnS (solid)}$$
$$\Delta G^\circ_{298} = -69{,}270 \text{ cal./mole.}$$

It is hoped that full details of this work on zinc and tin sulphides will eventually be published.

Dr. **O. Kubaschewski** (National Physical Laboratory): There is one minor point to be clarified which has some bearing on the evaluation of experimental thermochemical data but does not affect the practical significance of the results with which this discussion is mainly concerned. Dr. Bookey's results on equilibria involving calcium phosphates lead to the figures of his equation (7), which deviate considerably from those estimated by Dr. Richardson (referred to in the paper). Although the heat and entropy terms of the two equations differ widely, they give approximately the same free-energy values at the temperature of measurement. Considering the heat of formation of iron phosphide, I would say that Dr. Bookey's heat term of 58,500 cal. is highly improbable. The method of deriving heats and entropies of reaction from the dependence on temperature of measured equilibrium values is generally subject to error. It is certainly safer and generally more correct to accept only the free-energy values at the temperature of measurement and to combine these with either the heat or the entropy of the reaction, obtained from calorimetric measurements or by estimation. The paper by Tombs and Welch contains just such a careful discussion of experimental data, and thus illustrates the point I am trying to make.

Dr. **J. Pearson**: Mr. Brandt's remarks should not be taken as a refutation but as a continuation of my own. I did not say—and if I gave that impression I apologize—that the way the oxygen was introduced into the system, either as ore or gas, top-blown or bottom-blown, was solely responsible for prior removal of carbon or phosphorus. Mr. Brandt's statements indicate how the method of working may affect the type of slag formed

and how this in turn can favour one reaction at the expense of the other. What Dr. Colclough said does not contradict what I wished to say, which was that the indications of a thermodynamic treatment of equilibrium systems, which would predict, for example, that it is not possible to remove phosphorus in appreciable amount before carbon has been brought to a very low level, should be viewed with reserve because kinetic factors—the type of slag and the way in which the furnace is worked—may be more important.

CORRESPONDENCE

Mr. **C. E. A. Shanahan** (British Iron and Steel Research Association) wrote: The removal of phosphorus from molten iron and steel involves the use of oxygen and basic materials, both of which should be as highly active as possible; this is one of the reasons why the basic open-hearth refining period is considered to be the best time for phosphorus removal. However, many basic open-hearth plants, especially those not possessing tilting furnaces, would welcome any method for reducing their phosphorus burden by some sort of hot-metal treatment, and the paper by Maddocks and Turkdogan is therefore of very great interest. The presence of high concentrations of carbon in hot metal severely limits the working potential of oxygen by the 'leak' reaction C/CO, and to combat this an extremely stable slag–phosphate compound is required. The authors have accordingly investigated the use of Na_2O, and under certain conditions they have obtained considerable dephosphorization. From the practical point of view it would be desirable to remove the phosphorus in the ladle or mixer, and this immediately limits the working temperatures to 1200–1400° C. and requires that the carbon content of the iron should be about 3% or more. Unfortunately, the amount of dephosphorization obtained by the authors at this level of carbon is very small. Thus, from Fig. 3 of their paper, $\frac{(\Sigma Fe)}{(\Sigma SiO_2)} \simeq 0$ for C $\simeq 2{\cdot}4\%$, and Fig. 4 shows the negligible amount of phosphorus removed. In contrast to this, a detailed study of mixer operational data has shown that some active mixers are capable of removing 20% of the phosphorus at a [P] level of about 1·5% in the presence of [C] > 3%, allowing for the decrease in [P] due to scrap dilution.

It would be of practical interest to know if there was any marked increase in the silicon content of the metal after the various dephosphorization treatments. If this was high, the use of the authors' treatment on a large scale might have to be accompanied by a desiliconization treatment, since the slag bulk in the basic open-hearth furnace tends to be a function of the *total* acid-forming constituents that are charged.

The behaviour of Na_2O-bearing slags on sulphur removal is of great topical interest, and it is hoped that extra information on this phase of the work will be forthcoming. In this connection it would be valuable to ascertain whether any of the sulphur has migrated to the furnace atmosphere. This could most easily be done by means of a sulphur balance performed at the beginning and the end of each heat. The weight of sulphur introduced initially to the system can be estimated and the weight of sulphur finally present can be calculated from the final metal and slag analyses and their weights, the slag weight being determined by a lime, manganese, or phosphorus balance. All the necessary data must be available, and therefore an estimate of the average percentage of the sulphur lost to the furnace atmosphere would form a useful addendum to the paper. Certain mixers produce considerable desulphurization of the hot metal, and in some cases the mixer atmosphere is responsible for a large proportion of the total sulphur removed.

A knowledge of the activity coefficients of sulphur in the presence of C, Si, etc., is of importance in the study of hot metal desulphurization, and, as pointed out by Richardson and Jeffes in their paper, the values obtained by the H_2/H_2S methods may be in error because of the formation of compounds such as CS. Extra information that might support the accuracy of any of the measured sulphur activity coefficients is therefore of interest. Samarin and Shvartsman[12] assumed that at 1600° C. the number of sites available for S or C in molten iron is equal to one quarter of the number of Fe atoms, and that only 1 atom per site is permissible at any instant. This leads to the equation

$$\gamma_S = \frac{1}{1 - 5(N_C + N_S)}$$

where N_C and N_S are the atom fractions of C and S, respectively, and γ_S is the activity coefficient of sulphur. This equation has been applied to the results of Morris and Buehl[13] (at 1600° C.), and gives values of γ_S that are in extremely good agreement with those measured, up to about 3% of carbon. At carbon levels > 3%, the sulphur activity coefficients calculated from the above equation are higher than the observed values, *i.e.*, in the reverse order of magnitude to that predicted by the presence of CS. However, only two results are quoted for carbon levels of over 3%. At temperatures above 1600° C., $(N_C + N_S) > 0{\cdot}2$ and the sites available for sulphur and carbon can approach one third of the number of iron atoms. The above equation is then inapplicable.

Although the agreement between certain calculated and derived sulphur activity coefficients is worth noting, the above equation, by its symmetry, states that C and S are replaceable atom for atom without influencing γ_S. It is doubtful whether this is generally true, particularly in view of the probable difference in structure between solutions of sulphur in liquid iron and carbon in liquid iron. Thus, as pointed out by Kitchener, Bockris, and Spratt,[14] carbon is the only element which would be expected to dissolve in molten iron with relatively little alteration of the Fe structure.

It is usual, in the combustion method for determining sulphur in steel, to renew frequently the filter plug inserted near the exit of the combustion tube. This is done to prevent the collected iron-oxide dust from either absorbing some of the sulphur or perhaps catalysing the formation of SO_3. In this connection, it would be of interest to know whether Fincham and Richardson's CO_2-method results were obtained on a combustion train already contaminated with iron oxide. If this were so the values would show that the dust does not absorb significant amounts of sulphur as SO_2, although nothing would be learned about its catalytic effect in SO_3 formation, because of the accompanying very low oxygen potential even in the cold part of the apparatus.

The values which these authors obtained by combusting in pure oxygen are considerably lower than the more accurate gravimetric results. However, the technique employed differs in several respects from slag combustion methods used in practice. Nearly all the latter methods employ fuels such as iron or steel millings, which are inserted into the combustion boat together with the slag sample. During combustion, the overall temperature of the sample is considerably in excess of that of the furnace tube, owing to iron oxidation, and this is beneficial both in reducing SO_3 formation and in decreasing the total combustion time required. The combustion of the iron millings will tend to cause a low oxygen pressure in the vicinity of the boat, and this

again will retard SO_3 formation. A third retarding influence is provided by using gas flows in excess of that used by the authors, usually over 1 litre/min. (for similar bore tubes); this decreases the time for cooling the gases to 1000° C. and below. Naturally, high flow rates may result in incomplete absorption of the SO_3; the efficiency of the process should always be tested by adding further absorption vessels to the combustion train.

AUTHORS' REPLIES

Dr. **E. T. Turkdogan** (British Iron and Steel Research Association) replied: At the time this research was undertaken, it was questioned whether dephosphorization of steel could be brought about in the presence of high proportions of carbon and at temperatures lower than those in basic Bessemer and open-hearth processes. Before commencing the investigation some consideration was given to the factors which control the removal of carbon and phosphorus in steelmaking furnaces; as similar points have been raised in this discussion by a number of speakers, further thought should be given to these important facts. Although the removal of carbon is accomplished by its oxidation to a gas, in the case of phosphorus the oxidation product, being insoluble in steel, must be taken up by the slag. From the physicochemical viewpoint, dissolution of phosphorus pentoxide in the slag immediately results in a lowering of its activity, the reaction and its equilibrium constant being represented as follows:

$$2P + 5O \rightleftharpoons P_2O_5(\text{liq}). \quad \ldots\ldots\ldots\ldots\ldots\ldots (a)$$

$$K_P = \frac{(a_{P_2O_5})}{[a_P]^2[a_O]^5} \quad \ldots\ldots\ldots\ldots\ldots\ldots (b)$$

The sole reason for greater phosphorus removal under highly basic slags, for a given oxygen potential in the metal, lies in the lowering by basic oxides of the activity or activity coefficient of phosphorus pentoxide in the slag. Fundamentally, this is the result of tendencies towards compound formations in the molten slags. Therefore, the more stable the phosphate the less will be the activity of PO_4''' ions in molten slags, and for a given final [P] the $_{\mu}O_2$ in the iron, *i.e.*, oxygen potential of the melt, must also be lower. Assuming a C/O equilibrium in the melt, the above conditions will increase the equilibrium carbon concentrations in iron. In practice, there is never C/O equilibrium in steelmaking processes, and at the time of tapping the concentration of carbon in steel is about 10–20 times greater than the equilibrium conditions would justify. This is naturally attributable to the slowness of the C/O reaction compared with the transfer of oxygen from slag to metal or gas to metal. The results on soda-slag melts prove the above considerations.

At this stage I would like to elucidate Dr. Pearson's remarks on the selective oxidation of carbon and phosphorus, which appears to have been misinterpreted by Mr. Brandt. Whether oxygen-enriched or not, top-blowing in a basic converter with consequent formation of a highly basic slag reduces the PO_4''' activity to such an extent that, apart from equilibrium requirements, the driving force (*i.e.*, the rate of reaction) for the oxidation of phosphorus by equation (*a*) will be increased.

According to Mr. Shanahan's statement, some active mixers are capable of removing 20% of the phosphorus from a 1·5% P charge in the presence of 3% C. In none of our experimental melts could the carbon content be kept as high as 3%. This was mainly because of the technical difficulties involved in melting 600 g. of pig iron in an electric-arc furnace without some loss of carbon during the melting-down period. If these melts were to be carried out on a larger scale, the carbon level would no doubt have been raised. It would have been of great value to us if Mr. Shanahan could have included in his contribution some metal and slag analyses for these active mixers. In regard to the return of silica from slag to metal due to successive dephosphorization treatments by soda-slags or similar slags, I do not think that there will be any noticeable silica reversals, particularly when more basic oxides are used; this is because the activity of oxy-acid radicals as a whole will be reduced with the so-called basicity of the slag. Although definite relationships could not be shown with respect to desulphurization under soda-slags, it became clear that the extent of sulphur removal was much greater under these slags compared with those of the lime–silica type of similar acidity and oxygen potential. This would also indicate that Na^+ in slag reduces S'' activity to a greater extent than does Ca^{++}.

Mr. Evans' misgivings as to the work on soda-slags cannot be regarded as justified. His objections, as I understand them, are mainly concerned with the corrosive effect of soda-slags on steelmaking refractories. I should, however, point out that in choosing a refractory material for the lining of a furnace the first consideration is always given to the working temperature and the composition of the slag that the furnace will hold. Bearing in mind the important significance of soda-slags, would it not therefore be advisable to attempt to find a suitable refractory lining, rather than to ignore the usefulness of the soda-slag process?

It is well known that, in the past, many investigations have been carried out in industry and on a laboratory scale on refractories that will hold various types of slags. I see no reason why similar work should not be undertaken with respect to soda-slags. It may be worth noting that the experiments carried out on magnesia hearths were very satisfactory. Moreover, it should be borne in mind that these slags are meant to be employed at temperatures lower than 1400° C., and, as pointed out by Dr. Giedroyc, corrosive effect is less serious in mixer practice.

Dr. **F. D. Richardson** (Nuffield Research Group in Extraction Metallurgy) replied: The general points made by Mr. Evans and Mr. Oliver merit the serious consideration of all those concerned with research in extraction metallurgy. There is a gap, as everyone recognizes, between fundamental knowledge and works practice. This has its origin in the gap between those who conduct physicochemical research on metallurgical problems in the laboratory and those who make metals in the works. This gap can be filled by the research departments of works or by development groups in research associations. *It will not be filled satisfactorily, however, until the metallurgists and chemists in such groups have a reasonable understanding of elementary chemical thermodynamics.* This they rarely possess, in spite of the fact that there is nothing recherché about such knowledge. Fortunately, metallurgists in some Universities are now receiving a better training in chemical thermodynamics than hitherto, so there is some hope that this gap may be filled by the graduates of the next five years. There is no other way round this problem, for chemical thermodynamics is as important to the extraction metallurgists as are the concepts of elements and compounds or atoms and molecules.

Thermodynamic studies are made to determine the limits to which reactions can proceed. To the metal maker, they represent the more important of the twin approaches—thermodynamics and kinetics. The only adequate way in which the results of thermodynamic studies can be represented is in terms of heats, entropies, free energies, and activities, commonly in the form of

free-energy maps. Anyone who has to design or adapt extraction or refining processes should be able to read such maps if he is not to get lost in the world of molten metals and slags in which he works. No serious hiker would think of travelling without as good a map as he could buy, or would regard a knowledge of map reading as anything other than essential. When the men in the gap—the middle men—become able to think about works problems in a thermodynamic way, they will begin to do their job properly. They will begin to make use of the ever-growing body of basic knowledge of simple reactions and, as a result, they will solve their complex problems more quickly and at less cost than hitherto. Until this happens, the results of fundamental research will lead to nothing more than clearer academic thought and a novel process here or there. They will have little or no influence on current methods of production or current costs.

Mr. Oliver has referred to the division of FeO in steelmaking slags into 'free' and 'combined' oxide, and has asked how this division should now be regarded. Dr. Pearson has already partly answered this question. I would merely add that the thermodynamic way of dealing with it avoids the absurdities which can follow from the old division of FeO in this way—the absurdity of sometimes requiring that, in a slag containing, say, 40 mol.-% FeO, 50% should be free and − 10% combined! The FeO is now assigned an activity coefficient (γ_{FeO}) which is dependent on slag composition and may be greater or less than unity. This activity coefficient has to be measured for a number of slag compositions, and can then be estimated for other slag compositions, as *it varies continuously and not discontinuously with composition*. The activity (a_{FeO}) that governs the thermodynamic (or equilibrium) behaviour of FeO in such slags is then given by the equation

$$a_{FeO} = \gamma_{FeO} \times N_{FeO},$$

where N_{FeO} is the molar fraction or mole per cent.

Both Mr. Evans and Mr. Oliver have raised queries about SiO. Now that the thermodynamic properties of SiO are reasonably well established, it is possible to give them quantitative answers so far as SiO *gas* is concerned. The pressure of SiO which is in equilibrium with a silica roof and incompletely burnt gases can be quite large—at 1700° C. it would be about 1 mm. at a CO/CO_2 ratio of 100/1, and about 0·01 mm. at a ratio of 1/1. It is clear, therefore, that a silica roof could be rapidly converted to vapour, and so worn away, under conditions where large volumes of unburnt gas come into contact with it. The pressure of SiO gas in equilibrium with the slag system referred to by Mr. Oliver is only about 0·01 mm., so that SiO *gas* plays no rôle under such conditions. This does not mean that there cannot be SiO dissolved in the slag under the same conditions, for there is no precise information on its solubility. We would not, however, expect it to be significant at such low SiO pressures and, in addition, the analysis of oxidizing slags does not suggest any marked divergence from the stoichiometric composition $FeO–SiO_2–Fe_2O_3–CaO$, etc. The SiO might conceivably become important in slags made under reducing conditions—*e.g.*, in the blast-furnace. There is, indeed, evidence to show that there is too little oxygen in these slags, owing to the presence either of dissolved SiO or of dissolved or suspended SiC.

Aluminium might, as Dr. Leckie suggests, be a possible desulphurizer, but it does not look promising. The point for $Al–Al_2S_3$ in our sulphide diagram is not too definite, but the $Al–Al_2S_3$ line must slope upwards, and, at 1600° C., $\Delta\mu_{S_2}$ might be about − 80 kg.cal. This is about the value for sulphur in molten iron at 0·01%. The activity coefficient of aluminium in iron has, however, the rather low value of about 0·025.[15] One would therefore not expect 1% of Al in iron and free Al_2S_3 to be in equilibrium with less than 1 or 2% of sulphur in the metal. Although, at first sight, aluminium looks a better desulphurizer than manganese, its effect is spoilt by its much greater affinity for iron, which is the cause of its low activity coefficient.

The CS line is not accurately known but, if the line shown in our paper is approximately correct, carbon cannot be a reasonable desulphurizer; CS could only be really useful under conditions where its line falls *well* below that for [S] at 0·1% in iron (not shown on our diagram). It is not a question of the intersection of the Fe–FeS and C–CS lines. The hypothetical temperature for the equilibrium $C–CS_{(1\ atm.)}–[S]_{(0\cdot1\%)}$ (*i.e.*, the intersection of the lines for CS and $[S]_{(0.1\%)}$) would be 3300° C.

Dr. Leckie has also asked whether the phosphate in basic slags should be considered as tri- or tetraphosphate. From a thermodynamic standpoint it does not matter how it is regarded, but in calculations on such slags I think that it would be simpler to work in terms of triphosphate activities rather than tetraphosphate activities. The difference in $\Delta G°_{formation}$ is only about 3 kg.cal. at 1600° C., so that the value of $a_{Ca_3(PO_4)_2}$ will be greater than that of $a_{Ca_4P_2O_9}$ under conditions where a_{CaO} is less than about 0·45.

I agree in general with the remarks of Dr. Colclough, Mr. Brandt, and the Chairman, concerning the effect of slag basicity on the order of removal of phosphorus and carbon. I consider that, at present, there is insufficient evidence for the kinetic explanation suggested by Dr. Pearson. Figure 7 in the paper by Bookey, Richardson, and Welch shows that the carbon and phosphorus deoxidation lines cut at about 0·1% [P]. For a very basic slag containing small amounts of phosphate, the phosphorus line will be lowered, so that phosphorus will be removed in preference to carbon as the bath is oxidized. For a weakly basic slag the line will be raised, and carbon will tend to be removed in preference to phosphorus. The silicon deoxidation line lies further below that for phosphorus; furthermore, it is markedly lowered if the silicon is converted to silicate, as it would be in a basic slag, and not to silica. It is thus certain that silicon removal must precede phosphorus removal in any event. This point has been referred to by Mr. Evans. I should perhaps stress that it has nothing to do with silico-phosphate formation in the slags.

I agree with Dr. Richards and thank him both for his comments and for the valuable additional sulphide data he has given us. I am also in full agreement with Dr. Kubaschewski's comments concerning heats and entropies.

I do not think that the equation for $\gamma_{[S]}$, quoted by Mr. Shanahan and derived by Samarin and Shvartsman from a most elementary model of solutions of sulphur and carbon in molten iron, can be of any value at all in checking the accuracy of our CS equation. This has to be checked by reference to something likely to be more accurate, not less accurate, than itself.

In Fincham and Richardson's CO_2-method of sulphur analysis, a few steels were analysed in the combustion apparatus, so some slight contamination of the train by iron might have occurred. As Mr. Shanahan suggests, this obviously did not absorb any SO_2. I agree with him that our combustion technique in oxygen differed from the usual in the ways he mentions. In my view, the best way of decreasing the amount of SO_3 formed in the oxygen combustion technique would be to cool the gases as quickly as possible after they have passed the slag. By means of a rapid quench effected by rapid flow of the gases down a narrow tube, it might

be possible to decrease the SO_3 loss to much smaller proportions. This would seem to be worth attempting.

Mr. **A. Jackson** (Appleby-Frodingham Steel Company) replied: Mr. Cameron discussed the question of sulphur elimination with high slag volumes. We have already given consideration to the possible effect on sulphur removal of the differing sulphur contents in the gas, without finding any positive correlation. We cannot go back to the old practice of using only blast-furnace gas, which is sulphur free, as with a gas of such low calorific value we cannot heat the furnace sufficiently to enable high slag volumes to be used.

In Table V of the paper, the comparative effects of feeding a mixer quickly and not feeding it at all are shown. These two mixers were using the same type of fuel, *i.e.*, cold coke-oven and blast-furnace gas. One mixer produced 320 tons of slag per week and the other 30 tons. A short calculation will show that, with low slag volume, important quantities of sulphur have again disappeared from the bath, and this appears to indicate that the particular phenomenon is not related, in any direct way, to the sulphur content of the atmosphere above the furnace bath.

It is difficult to give Mr. Walker a direct reply to his enquiry as to the most desirable manganese content in the iron. Figure 8 of the paper shows two curves relating the manganese content of the iron to the sulphur as it leaves the blast-furnace. This indicates that blast-furnace operation (probably lime/silica ratio) has an important bearing on the ratio of manganese to sulphur. Probably Mr. Walker had in mind the effect of manganese in the iron at the melting shop. A study of the problem has indicated that, with a given manganese content in the iron, the amount of sulphur eliminated in the active mixer decreases with increased rates of feeding. If the manganese content of the iron is increased, a curve of similar form is produced but at a higher level, *i.e.*, giving more sulphur removal for the same rate of feeding. In other words, with a given rate of feeding the amount of sulphur removed from the mixer iron increases with increasing manganese content in the incoming metal.

In my opinion, for iron of the quality we use, a manganese content of 1·25–1·5% is advantageous.

We have considered Dr. Leckie's suggestion of desiliconizing with scale and speeding up the reaction by blowing oxygen as well. We have not tried this, for reasons mentioned in the paper, *i.e.*, the question of fumes, transfer ladles with limited capacity, and so on. There appears to be little doubt that this process is advantageous, but it is not specifically applicable under our conditions at the present time.

Dr. Leckie also remarked on the reduction of phosphorus in the mixers and the possible value that might arise from an investigation of the slag analysis by those interested. If anyone wishes to have samples of individual mixer slags under various conditions we would be very pleased to supply them. They would probably materially assist theoretical work by showing what actually occurs in practice.

It was most interesting to hear from Mr. Brandt of the excellent work being done at Brymbo. He states that sulphur is removed; does he know whether the sulphur has already been removed between blast-furnace and his oxygen-blowing process by flotation, or whether it was actually lost during the course of blowing in contact with lime?

H. Flood and **K. Grjotheim** (Institute of Silicate Science, Norway) wrote: Some corrections in our paper are necessary. On p. 65, reaction (1) should be written

$$\Delta\mu_1 = \mu_{(Na,K)Br} - \mu_{(Na,K)Cl} + \tfrac{1}{2}(\mu^\circ_{Cl_2} - (\mu^\circ_{Br_2})$$

and reaction (3) should be written

$$\Delta\mu_3 = N_{Na^+}(\mu^\circ_{NaCl} - \mu^\circ_{NaBr}) + N_{K^+}(\mu^\circ_{KCl} - \mu^\circ_{KBr}) - \tfrac{1}{2}(\mu^\circ_{Cl_2} - \mu^\circ_{Br_2}).$$

On p. 66, the formula for reaction (1) should read:

Reaction (1)—N moles of Fe (A) are transformed reversibly into Mn (A):

$$\Delta G_1 = -NR'T\log K = -NR'T\log K' + f(\gamma),$$

and on p. 68, in the formula at the foot of the right-hand column, each N should be replaced by N'.

Some recently published data of Bell, Murad, and Carter[16] on the manganese distribution between iron bath and basic slags were used to demonstrate the practical application of the additivity equation when the constants of the pure systems are known. The analytically determined manganese content of the iron phase was compared with the manganese content as calculated from the slag analysis using the additivity formula (*see* Table *B*). The agreement between them is fairly good, except for Heat No. 20, for which Bell and his co-workers confirmed that their measurement was wrong—the amount of trivalent iron was not determined, and the influence of this iron was therefore neglected in the calculation. However, the influence of Fe^{III} will be significant only when $N'_{O^{2-}}$ is close to 1. Similarly, at $N'_{SiO_4^{4-}}$ close to 1, the influence of $Si_2O_7^{6-}$ ions will probably be noticeable.

The concept of the iron oxide activity in a basic slag is dealt with in a paper which is shortly to be published.[17] In this paper a comparison is also made between Dr. Bookey's constants and ours.

Table B

CALCULATED VALUES OF MANGANESE IN IRON COMPARED WITH EXPERIMENTAL VALUES OF BELL, MURAD, AND CARTER, AT 1550° C.

Heat No.	Slag Analysis. Ionic Fractions					Manganese in Metal, wt.-%	
	$N_{Fe^{2+}}$	$N_{Mn^{2+}}$	$N_{Mg^{2+}}$	$N_{O^{2-}}$	$N_{SiO_4^{4-}}$	Experimental	Calculated
11	0·780	0·158	0·072	0·952	0·048	0·050	0·053
13	0·704	0·123	0·173	0·901	0·099	0·04	0·04
14	0·552	0·285	0·163	0·912	0·088	0·12	0·12
15	0·286	0·619	0·095	0·876	0·124	0·46	0·48
16	0·302	0·605	0·093	0·773	0·227	0·40	0·36
17	0·426	0·474	0·100	0·793	0·207	0·20	0·20
18	0·373	0·504	0·123	0·715	0·285	0·22	0·22
19	0·663	0·129	0·208	0·795	0·205	0·032	0·035
20	0·449	0·352	0·199	0·637	0·363	0·085	0·107

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Discussion on the Paper—

ADDITION OF BORON TO STEEL BY REDUCTION FROM BORON OXIDE*

By G. E. Speight

Mr. **G. E. Speight** (United Steel Companies, Ltd.), in presenting the paper, drew attention to an error in the third complete paragraph of the first column on p. 153, where the formula for sodium borate was given as $Na_2B_2O_7$, whereas it should be $Na_2B_4O_7$. That error also altered the calculated yield of boron given in the following paragraph, halving the figures.

Mr. **D. A. Oliver** (B.S.A. Group Research Centre): Some very important results have been obtained following a chance observation of the apparent contamination of a steel by a minute amount of boron picked up from a furnace lining. In my view, this work has emphasized the right factors, with the important conception of acid-soluble and insoluble boron and the determination of which variety is effective in increasing hardenability and which is not. This is going to advance the subject fundamentally.

The author was very cautious in referring to what he calls the 'metallic boron,' which is the effective acid-soluble boron and may well be in solid solution. Does he agree that any metallic boron present would readily form boron carbide, or perhaps go into solid solution in the other carbides present? Might it not also form intermetallics with the iron base? I think it is improbable that it would be present without some of its chemical affinities being satisfied.

I found difficulty in linking up the description of the results given in Tables II and III with Fig. 1. The description in the text agrees with Table II, but the hardenability curves in Fig. 1 appear to have been labelled wrongly. The curve is displaced to the right, which shows an appreciable increase of hardenability, but it is described as being insignificantly improved from the control, which was, of course, boron-free. If the curve with crosses, on the extreme left, is the control curve, it would seem that *H*6770 should be the one near the control, and the one to the extreme right should be *H*6771. Perhaps the author will clarify this point?

With regard to the author's recommended deoxidation practices, in general I am strongly against deoxidizing any steel with aluminium if it is possible to avoid it. In a killed steel, aluminium deoxidation, in my experience, usually produces massive inclusions in odd places in an ingot, leaving the rest of the steel abnormally clean. As a rule, a very fine dispersion of unavoidable oxides or non-metallic inclusions is preferable to a very clean steel with a few massive inclusions. That line of thought caused me some years ago to follow up the potentialities of deoxidizing with silicon–zirconium, although zirconium is not quite so powerful as aluminium.

I see no objection to the addition of 1 lb. of aluminium per ton before the slag treatment (p. 133), but I was amazed at the further addition of 2 lb. of aluminium per ton, and finally the relatively large amount of ferro-titanium. There would appear to be a risk of forming what is usually called 'titanium cyanonitride,' *i.e.*, a solid solution of titanium carbide and titanium nitride in cubic pink crystals. Has the author experienced this?

In view of the foregoing, what does the author think of the following procedure: initial deoxidation with ½ lb. of aluminium per ton, mixed with or simultaneously added with the equivalent of ½ lb. of calcium per ton in the form of calcium–silicon–manganese? That combination of aluminium and CaSiMn I have previously found very effective. The calcium cleanses the steel and acts as a grain-coarsener, while the aluminium acts as a grain-refiner. The two apparently fight, but in general the aluminium wins, and a thoroughly clean steel results. Finally, instead of titanium I should prefer to add silico-zirconium, because the zirconium will fix the nitrogen and certainly stabilize the steel against reoxidation, and the inclusions, if any, will be finely dispersed. Moreover, if any glassy silicates survive this ordeal the zirconium will tend to attack them and turn them into smaller dispersed particles.

Much publicity has been given to boron in relation to alloy economy, particularly in American literature. Under certain conditions boron additions can have real advantages, but I think the subject has been over-sold. The author has clearly shown how it can be tackled scientifically and real advances made.

During the 1939–45 war, when the special boron addition agents were first introduced to Great Britain,

* *J. Iron Steel Inst.*, 1952, vol. 171, June, pp. 147–153.

Table A—EFFECT OF BORON

Quality	Ni-Mo			Ni-Mo-B			SAE. 4140	SAE. 1035 + Boron	0·45–0·55% C, 1·5–1·8% Mn		
Steel or Cast No.	1	2	3	4	5	6			SG7768	SG7769	SG7770
Analysis											
C, %	0·15	0·15	0·14	0·14	0·15	0·14	0·42	0·36	0·45	0·47	0·46
Mn, %	0·55	0·59	0·55	0·55	0·57	0·55	0·96	0·86	1·44	1·48	1·60
Si, %	0·28	0·28	0·27	0·30	0·30	0·30					
Ni, %	2·11	2·07	2·06	2·14	2·08	2·12					
Cr, %							0·98				
Mo, %	0·12	0·21	0·30	0·14	0·21	0·31	0·22		Nil	0·0027	0·0029
B [sol.], %				0·0014	0·0017	0·0016		0·0023			
Mode of B addition										Silicaz	Al, ferro-boron
Size of material										1½-in. dia. bar	
Heat-treatment	Blank carburize 920° C., 4 hr. F.C., refine 860° C., O.Q., harden 780° C., O.Q.						¾-in. dia. bar O.Q. 860° C., T. 640° C.	¾-in. dia. bar O.Q. 860° C., T. 500° C.	O.Q. 860° C., T. 550° C.	O.Q. 860° C., T. 590° C.	O.Q. 860° C., T. 590° C.
Mechanical Properties											
U.T.S., tons/sq. in.	46·9	52·0	52·6	51·3	59·5	61·4	67·3	66·6	59·2	58·7	60·0
Yield stress, tons/sq. in.	32·3	31·8	38·7	37·4	42·3	42·8					
0·2% proof stress, tons/sq. in.							59·5	60·9	42·0	49·4	50·4
Elongation on $4\sqrt{A}$, %	35·5	29·5	27·0	31·0	25·0	23·5	22·0	19·5	18·0	20·0	20·0
Reduction in area, %	59·6	51·9	52·5	55·4	51·9	50·4	54·9	56·7	41·8	52·0	51·0
Izod impact, ft.lb.	63	53	51	57	48	47	48	42	11	37	33

they seemed to be effective in only very few cases. They were notably successful in En 16, a Mn–Mo steel. Increased proof stresses were obtained in straight Cr steels when lightly tempered; but for most other steels they conferred no observed advantages. I think this was because boron was added to steels that were already quite highly alloyed. The Americans start virtually with a pure carbon steel, that is, one remarkably free from tramp elements. In Britain it is difficult with the usual carbon scrap to keep below 0·3% of Ni. If, then, the Americans add a small amount of some alloying constituent such as 0·5% of Cr and then take hardenability curves, they definitely find that small boron additions in most cases give better depths of hardening, or hardenability. On that basis, coupled with the fact that they can then class it as an alloy steel, they claim that they are getting substantial advantages.

It may be that metallurgists in Britain are just moving into the phase when boron can be used intelligently, as distinct from merely being added without a very clear idea of its advantages. As Britain is progressively forced to use steels with less and less alloy, a controlled addition of boron will probably prove most helpful.

Dr. **W. Sachs** (London and Scandinavian Metallurgical Co., Ltd.): One of the great problems in using boron in steel, and also in making boron alloys for the steelworks, is the difficulty of analysis. It is really most remarkable that in this work not only is 0·0001% of boron analysed, but this figure is even divided into two distinct parts.

Mr. Speight differentiates between soluble and insoluble boron, but he obviously cannot differentiate analytically between the soluble boron present as oxide and that in metallic form. For this purpose he depends on indirect methods, and I would ask him to confirm that. Is the usual colorimetric method employed for the determination of both insoluble and soluble boron?

To avoid the difficulty of boron analyses in the steelworks, the Americans recommend that the free nitrogen content of the boron-treated steel should be determined. However, this figure would not indicate the presence of free boron in soluble form, which according to Mr. Speight's findings is all-important.

On page 149 of his paper, Mr. Speight assumes that, in line with the theory, Al goes for the oxygen and the Ti and Zr for the nitrogen; does this reaction really take place in this way?

In the experiment described on page 151, Mr. Speight works with a more or less oxidic slag, and he gets some boron into the slag and some into the metal. Afterwards, he finds that the boron in the metal is present as oxide. Is, therefore, the boron properly reduced into the metal in the first instance and then oxidized again in the metal by residual oxygen? Or is there a direct solubility of boron oxide in steel?

With regard to the question of boron *versus* Al as a deoxidizer (page 150), I believe boron must be a weaker deoxidizer than Al, because boron alloys are made by reduction of boron oxide with Al.

In the manufacture of complex boron alloys a yield of 70–80% is obtained for boron and Ti—always, of course, in the presence of excess Al. The yield of Zr is much less, and the yield of Mn and vanadium, if added, is much greater.

The use of electric furnaces to produce boron alloys is convenient, but not essential. The capacities here and in the U.S.A. are sufficient to cover any likely requirements.

Finally, can Mr. Speight shed some light on the connection between the high-tensile and proof strengths which boron confers on Fortiweld steel, and the general hardenability question as raised by the Americans?

Mr. **W. E. Bardgett** (The United Steel Companies, Ltd.): Our experience in the production of boron-treated steels has been confined mainly to the low-carbon, ½% Mo steel, Fortiweld, made in the high-frequency furnace, in the arc furnace, and in the open-hearth furnace by both cold- and hot-metal processes. The effect of boron on this steel is to increase the ultimate tensile stress and proof stress in the as-rolled or normalized conditions. This is a special case and is probably the only exception to the general rule that boron is added to low-alloy steels to increase hardenability, with its consequent effect on mechanical properties. Whatever the purpose of the boron addition may be, the technique of its addition is the same.

Some earlier views on the effect of boron on hardenable steels may require modifications in the light of new knowledge and experience. For instance, the view had been held by some that boron treatment presents difficulties in control which are not experienced with the usual alloying elements. Our investigations on boron additions were made initially in an 18-lb. high-frequency furnace, and Mr. Speight built up, with complete success, from that experience to other steelmaking processes. Although boron cannot be added satisfactorily simply as a ferro-alloy, there is no 'hit or miss' about it provided that the necessary known precautions are taken to ensure that the boron is in the effective form. With experience it is possible to obtain really surprising control over the boron content, in spite of the very small amount required.

Certain variable and erratic results were obtained in earlier trials and these may be explained partly by lack of control of the introduction of boron into the steel. A point affecting comparisons arises from recent American work, which has led to the conclusion that the advantages of boron are obtained only when comparisons are made on the basis of 100% martensite in the boron steel. This, of course, reduces the section size on which it was at one time thought possible to obtain advantages in boron steels.

Earlier work led to the important conclusion that boron treatment of any one grade of En steel would not upgrade that steel to the next higher grade, *e.g.*, En 16 to En 17. The effect of boron on American grades is difficult to assess, since there appear to be little data on steels with and without boron; rather, the boron steels appear to have been designed to give hardenabilities equivalent to those of steels of higher alloy content. Clearly, boron does increase hardenability of certain steels, but there are very distinct limitations. For example, in low-alloy steels the effect of boron decreases with increasing carbon, and with eutectoid composition the effect is nil.

A few results from our investigations on the effect of boron are given in Table *A*. All these steels were made in an 18-lb. high-frequency furnace and were forged to a suitable size bar.

The first group shows two series of En 34 type steels varying in Mo, the first containing no boron and the second with boron added. The specimens were all given the standard heat-treatment for En 34 quality. The boron addition increases the tensile strength progressively with increase in Mo, the ductility and toughness being correspondingly reduced. Equivalent mechanical properties are obtained in a boron steel with about 0·15% of Mo and in a boron-free steel with about 0·25% of Mo.

The experience in America has not been altogether successful with carburizing steels. Unexpected difficulties have arisen owing to pronounced distortion and to the different effects of boron on the high-carbon case and the low-carbon core.

The next group shows two American steels, SAE.4140 and a boron-treated plain carbon steel, SAE.1035, that is stated to replace SAE.4140. The mechanical properties in the quenched and tempered conditions agree very closely, although the boron steel requires to be tempered at 500° C., compared with 640° C. for the higher-alloy steel. The hardenability curves for these two steels (Fig. *A*) show that there is no comparison at all on the basis of hardenability. All that can be said is that the boron steel gives adequate properties in the size under consideration. This is put forward as a warning that care should be exercised in interpreting claims made for boron additions.

The third group deals with a carbon–manganese steel required to give a minimum proof stress of 45 tons/sq. in. in the form of 1½-in. dia. bar. At the bottom of the specified range of 0·45–0·55% C, 1·50–1·80% Mn, there is difficulty in meeting the specification, and at the top there are other difficulties. A steel at the top of the range cracked on quenching and could not be used for the tests. *SG*7768 is close to the bottom of the analysis range; *SG*7769 and *SG*7770 are of similar composition, containing boron added as Silcaz and as aluminium and ferro-boron, respectively. The steels were all treated to a similar tensile strength, the boron steels requiring a higher tempering temperature than the boron-free steel. The results show that the boron-free steel does not meet the requirement of 45 tons/sq. in. minimum proof stress, whereas the boron-treated steels are satisfactory and, in addition, are very much superior in ductility and toughness. The hardenability curves for the three steels (Fig. *B*) clearly show the pronounced effect of boron; part of the improvement in *SG*7770 will be due to the somewhat higher Mn content.

Although obtained from tests on material made in a small laboratory furnace, the results indicate that boron can improve mechanical properties. Although it may not be possible to upgrade a steel from one grade to the next, there may be applications where steel is required in bulk, and even a small benefit from a boron addition may justify its use.

Fig. A—Hardenability curves for steels SAE.4140 and SAE.1035 + boron (substandard ½-in. dia. specimens)

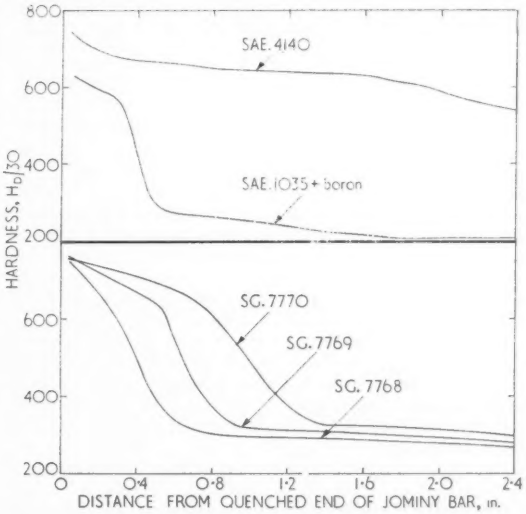

Fig. B—Hardenability curves for three carbon–manganese steels

Dr. **L. Reeve** (Appleby-Frodingham Steel Co.): At Appleby-Frodingham a fairly substantial tonnage has been made of the low-carbon molybdenum-boron steel, Fortiweld. Much of Mr. Speight's work provided the basis upon which production of this steel ultimately began.

We have discussed making this steel by the borax slag procedure, but have not yet used that method; the greater part of our steels have been made with ferro-boron and other deoxidizers added to the ladle. Complex deoxidizers have now been almost eliminated and, on the whole, it is preferred to work without them.

At the outset of production we were concerned about possible difficulty in getting the right boron content. The specification for this steel is 0·0015–0·0035% of soluble boron and our aim is 0·0025%. In 90% of a substantial tonnage the boron has been between 0·0020 and 0·0030%. Of the other 10%, the lowest was 0·0018% and the highest 0·0032%. Apart from some minor troubles at first, there has been no substantial difficulty in meeting the mechanical test specifications, which involve a very high yield point and proof stress in the as-rolled condition.

So far as is known, we are the only steelworks making a boron steel that can be used in either the normalized or the as-rolled condition. Most of the American work, and much discussion, is concerned with heat-treated steels, quenched and tempered; but, whilst Fortiweld can be quenched and tempered, it is normally used in the as-rolled condition. Since 1949 the impact properties of the steel as-rolled have been improved, or, more accurately, the transition temperature is now very much lower.

In a paper on these steels, published in 1949,* it was pointed out that, in the as-rolled condition, heavy plates, in particular, were likely to have very low impact values, implying that this steel would have to be used in the normalized condition. Since then, largely by changes in deoxidation procedure, that position has completely changed. On the main tonnage of material, the Izod impact has averaged about 70 ft.lb. in the as-rolled condition, and about 50, 40, and 30 ft.lb. on plates $\frac{3}{4}$–$1\frac{1}{2}$ in. thick.

Provided that the furnace melting procedure is reasonably standard and the deoxidation procedure is standardized very carefully, these molybdenum-boron steels can be made with reproducible properties from cast to cast without undue difficulty.

AUTHOR'S REPLY

Mr. **G. E. Speight** replied: I would like to thank those who have contributed to the discussion and to express my appreciation of the many kind remarks. The discussion has covered a wide field of boron ferrous metallurgy, and many important points have been made and new information brought forward that will doubtless be carefully studied by those interested in the uses and applications of boron steels. I propose, however, to confine my reply primarily to those questions relating to the chemistry and manufacture of boron steels.

In reply to Mr. Allison, there seems to be no reason why boron additions should not be made successfully to crucible steels; indeed, many hundreds of boron steel melts have been made in sillimanite crucibles in an 18-lb. H.F. furnace, which clearly has much in common with the crucible process.

Mr. Oliver has referred to the form of occurrence of the 'effective acid-soluble' boron. It seems hardly credible that the random distribution of such a small amount of boron (approx. 1 atom of boron to 12,000 atoms of iron) could have such a marked effect, unless there is some ordered orientation or distribution of boron in relation to the other constituents or to the general structure. It is possible that a complete and precise definition of the form of the effective boron will be forthcoming only when the mechanism of boron hardening is finally elucidated.

The differences in the three hardenability curves shown in Fig. 1 are slight and could easily arise from the inevitable minor differences in general composition. In fact, the boron-containing steel *H*6770 shows slightly better, and the boron-containing steel *H*6771 slightly poorer, hardenability than the control test, *S*8338. Steel of this quality (0·35% C, 1·5% Mn, 0·25% Mo) should show a marked increase in depth hardenability on being effectively boron-treated. This is illustrated by the following data:†

	Element, %			At Inflexion		
	C	Mn	Mo	$H_D/50$	Jominy Distance, in.	Ideal Dia., in.
Steel *D8A* (no boron)	0·32$_8$	1·35	0·32	400	0·49	2·9
Steel *D8B* (boron-treated)	0·33	1·33	0·33	415	1·0	4·6

Similar tests in our laboratories confirmed this order of improvement on addition of boron.

Mr. Oliver's aversion to aluminium treatment of killed steels is noted, and presumably this extends also to fine-grained steels. However, it would appear that boron requires a fine-grain environment to develop its full benefits, and that some degree of aluminium treatment is therefore essential. In regard to the amount of aluminium specified, it should be remembered that the recommendations on pp. 152–153 of my paper relate to a low-carbon steel and to the addition of boron from a borate or boron-containing slag. Both factors necessitate a greater aluminium treatment than would be required by a medium-carbon steel from the usual boron addition agents, *e.g.*, ferro-boron. Allowing for these factors, the recommended aluminium treatment is not very different from that which would be required for a normal fine-grain treatment.

Mr. Oliver's proposed alternative deoxidation scheme is interesting; it reveals a true appreciation of the factors involved in the success of boron additions, and the calcium–manganese–silicon should certainly help in eliminating deleterious aluminous inclusions. In conjunction with a normal boron addition agent, *i.e.*, ferro-boron, it should give reasonably satisfactory results for medium-carbon steels of normal nitrogen content (0·005%), provided that the boron addition is increased to allow for some loss of boron as the 'acid-insoluble' constituent. An increase in the aluminium addition would, however, provide a greater margin of safety.

Zirconium is, as stated in the paper, a satisfactory substitute for titanium for nitrogen stabilization, and trials have been made to confirm this. It should be remembered, however, that, even assuming complete parity between titanium and zirconium for this purpose, the zirconium necessary would be almost double the titanium required. The zirconium addition would thus be a large and costly item in the boron treatment.

Dr. Sachs mentioned several points on the analysis and identification of these small amounts of boron. After extended trials of various methods of analysis, the direct colorimetric quinalizarin method described by Rudolph and Flickinger‡ was adopted, and it has been

* W. E. Bardgett and L. Reeve, *J. Iron Steel Inst.*, 1949, vol. 163, pp. 277–294.

† J. Glen, Symposium on the Hardenability of Steel, *Spec. Rep. No.* 36, pp. 374–375, Table CIV: 1946, London, The Iron and Steel Institute.

‡G. A. Rudolph and L. C. Flickinger, *Steel*, 1943, vol. 112, p. 114.

used almost daily for the past 5–6 years. It is not an elegant procedure, and only after considerable experience is it possible to obtain accurate and reproducible analyses. However, the method has been extremely useful and without it the vast programme of chemical work on boron steels would have been seriously retarded. Spectrographic analyses for boron are also carried out, but these give no direct evidence of the chemical form of the boron.

As Dr. Sachs suggests, the distinction between the effective and non-effective forms of ' acid-soluble ' boron is an indirect one. Only in isolated instances, however, has there been evidence of this non-effective variety. It occurs when the steel has been imperfectly deoxidized, *i.e.*, when the aluminium treatment has been either omitted or inadequate, or when some abnormality in casting has produced an over-oxidized condition, *e.g.*, oxygen lancing of the ladle nozzle. When the steel contains sufficient deoxidizers, the ' acid-soluble ' boron is substantially effective, but when deficient in deoxidizers, the steel may show more than the critical limit of ' acid-soluble ' boron and yet not have the expected enhanced properties. From this and other evidence of the actual oxygen contents, and from theoretical reasoning, we have concluded that the non-effective variety of ' acid-soluble ' boron is boron oxide, probably in combination with other deoxidation oxides.

A steel which has been boron-treated according to the principles outlined in the paper should contain a low free-nitrogen content, *i.e.*, the ' acid-soluble ' nitrogen. This value will also include any aluminium nitride, but this should be low if titanium or zirconium have been used as primary nitride-stabilizers. Consequently, Dr. Sachs's comment that a free-nitrogen determination might be an indication of whether the boron treatment has been satisfactory may well be feasible, provided that the steel-making conditions are reasonably uniform. Although, for routine control purposes, the free-nitrogen content might be acceptable, it could never fully replace the determination of the ' acid-soluble ' boron, and it would be necessary first to establish from previous experience the critical limits of ' acid-soluble ' nitrogen. However, such indirect tests are useful; from spectrographic determinations of the aluminium, titanium, and total boron contents it is possible to conclude whether the boron is substantially ' acid-soluble ' and effective, and to pronounce on a particular cast without carrying out an actual determination of the ' acid-soluble ' boron.

Concerning the chemical behaviour of aluminium and titanium, it is, of course, merely convenience of thought to regard aluminium as the deoxidizer and titanium as the denitridizer. The two elements can, and undoubtedly do, to some extent react with both gases. The relative affinities and concentrations are such, however, that aluminium controls the reactive oxygen while titanium controls the reactive nitrogen and prevents formation of boron nitride. When titanium is absent, aluminium takes over the rôle of nitrogen stabilizer, but in this case a much larger excess of aluminium is required to suppress or limit the formation of the unwanted boron–nitrogen compound.

The transfer of boron from slag to metal can be represented by a dissociation reaction as follows:

$$B_2O_{3\,slag} \rightleftharpoons 2\,B_{metal} + 3\,O_{metal}$$

The forward reaction is promoted (*i.e.*, the boron in the metal is increased) by (*a*) higher B_2O_3 in the slag, and (*b*) lower oxygen content of the metal. In the absence of other deoxidizers, such as aluminium and silicon, boron and oxygen co-exist in equilibrium in solution in the melt, and, on cooling, boron oxide (or its compound with ferrous oxide) precipitates as a deoxidation product. The presence of aluminium reduces the reactive oxygen and so assists the transfer of boron to the metal, and on cooling also controls the reactive oxygen to lower levels than are necessary for the formation of boron oxide. Thus, as already mentioned, boron oxide (or its compound with other deoxidation oxides) arises in the final cooled ingot only when the residual aluminium is too low.

Dr. Sachs's remarks on the relative yields of various elements in the aluminothermic reaction are, I believe, in accord with the general theory. In regard to boron, however, the paper describes how these beneficial, minute amounts of boron can be added to steel by direct treatment with boron oxide, thus avoiding the necessity for a separate intermediate reduction of the oxide either in an electric furnace or by an aluminothermic reaction. The direct use of boron oxide must have economic advantages, but there may well be certain factors that preclude its adoption to the complete exclusion of the more conventional boron-addition alloys.

In reply to Dr. Sachs's final question on the connection between the use of boron in Fortiweld and on the general field of depth hardenability, there can be little doubt that in both cases the effect of the boron is to delay transformation, and this may be regarded as equivalent to an increase in cooling rate. Similar means for adding the boron can be used in both applications.

I wish to thank Mr. Bardgett and Dr. Reeve for their kind references. Their remarks on some experiences of boron steels and on the present position with Fortiweld are interesting and call for no further comment.

Discussion on the Paper—

THE RAPID SOFTENING OF COLD-DRAWN AUSTENITIC STAINLESS STEELS*

By F. A. Hodierne and C. E. Homer

Mr. **F. A. Hodierne** (Department of Development and Research, Tube Investments, Ltd.) presented the paper.

Mr. **J. I. Morley** (Brown-Firth Research Laboratories): The authors have shown that thin sections of some austenitic stainless steels can be softened by what amounts to a flash heating at 1150° C. The specimens were free from intergranular corrosion when tested as water-quenched from the softening temperature, but no sensitizing heat treatment at 650° C. was given before the corrosion test. It cannot be concluded, therefore, that steels softened in this way are free from intergranular corrosion after exposure in the range 500–800° C., as in welding or brazing.

In the case of the 18/8/Ti steel, which is commonly

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used for welding, this might be a serious limitation, because the carbide stabilization with titanium becomes less effective as the softening temperature increases.

Twenty years ago the steelworks laboratories best known to me carried out intergranular corrosion tests in acid copper sulphate on five different casts of 18/8/Ti steel. Softening at 1250° C. followed by 5 min. at 650° C. resulted in severe intergranular attack in steels that were fully stabilized with titanium. Two of the five casts suffered intergranular corrosion after softening at 1200° C. plus 5 min. at 650° C., but only one after softening at 1150° C. plus 5 min. at 650° C. When softened at 1050° C., all five casts withstood the test, even after ½ hr. at 650° C.

A temperature of 1050° C. was chosen at that time to give fairly rapid softening with the least risk of weld decay in service. This became the standard softening treatment for the steel.

The mechanism of this susceptibility to intergranular corrosion in titanium-bearing steels softened at 1150° C. and above was not then understood. It is now thought to be due to the partial solubility of titanium carbide in austenite at high temperatures. The carbon taken into solution is reprecipitated, not as titanium carbide but as metastable chromium carbide, when first reheated to 600° or 700° C. This is presumably because there are more chromium atoms than titanium atoms, and because carbon cannot diffuse quickly enough at these lower temperatures to form the stable titanium carbide. The result is chromium depletion in the grain boundaries and intergranular corrosion.

The authors recommend a temperature range of 1100–1150° C. for softening the 18/8/Ti steel. If these temperatures are not exceeded, there may be no serious trouble, but how far is it possible to keep within these limits by induction heating in practice? Would there not be a tendency to err on the high-temperature side to ensure full softening? Have the authors tested any welded joints in 18/8/Ti steel tubes softened by this method for susceptibility to intergranular corrosion?

No tensile tests are reported in the paper on material softened by induction heating. Are the proof stresses in any such tests as low as would be expected from the hardness results? The tensile test indicates the properties of the core material of the strip, whereas hardness tests, if carried out on the surfaces, can be influenced by skin-heating effects.

AUTHORS' REPLY

Dr. **C. E. Homer** (in reply): We much appreciate Mr. Morley's contribution, and we would like to acknowledge the help that we received from him and his colleagues in the early stages of the work. His point about applying a sensitizing treatment at 650° C. after the rapid softening treatment could be important in some instances, as shown by his account of the behaviour of the titanium-stabilized 18/8 steel when heated to temperatures higher than those normally recommended. We have done some work on this point by rapidly heating a number of specimens of this steel to temperatures up to 1250° C.; after sensitizing at 650° C. for ½ hr., the specimens were not subject to intercrystalline corrosion, even to the slightest extent. It is possible, however, in view of Mr. Morley's remarks, that other casts of steel might have behaved slightly differently, and this is a point that would need to be followed up in the further development of the work. We should expect, however, that with very short times of heating the trouble referred to would not occur.

The point has been made that the temperatures that are desirable may be exceeded in practice. We have not much experience of this; we have done some induction heating of tubes on a pilot-plant scale, and have had some difficulty in controlling the temperature, but more work would have to be done on the engineering side before we could say definitely what limits could be maintained.

We performed tensile tests on the three 18/8 steels after tubes in those materials had been softened by induction heating, and we found the tensile properties to be closely in agreement with the hardness results quoted in the paper. These figures were not obtained on the surface of the specimens, but on cross-sections. We have not actually tested welds in tubes that have been rapidly softened.

Discussion on the Paper—

EFFECT OF DIRECTION OF ROLLING, DIRECTION OF STRAINING, AND AGEING ON THE MECHANICAL PROPERTIES OF A MILD STEEL PLATE*

By Constance F. Tipper, M.A., Sc.D.

Dr. **Constance F. Tipper** (University of Cambridge), in presenting her paper, pointed out that in the lower half of Fig. 1 of the paper as printed, the crosses should refer to the *Lx* direction and the circles to the *Xl* direction, and that in Fig. 2*a*, curve 5 should refer to the *Lx* direction. She then read a written contribution to the discussion by Mr. A. Cracknell and Dr. N. J. Petch (*see* p. 282).

Mr. **W. Barr** (Colvilles, Ltd.): On p. 145, discussing the Izod test results, Dr. Tipper states:

> "The first effect of the pre-strain is to lower all the values of energy absorption, particularly in the transition range."

The results of many notch-bar tests that I have carried out recently at room temperature on a particularly tough type of mild steel have shown that the effect of pre-strain in amounts up to 10% is very small, the reduction in energy absorbed being about 5%. This steel broke with a shear or fibrous fracture both before and after pre-straining; *i.e.*, pre-straining had raised the transition temperature, as is to be expected, but it had not raised it above the temperature of testing (room temperature). On the face of the evidence in Fig. 2 of the paper, it is difficult to understand why Dr. Tipper did not describe the effect of pre-strain in the same terms, *i.e.*, its effect on the transition temperature, which is acknowledged to be a rational criterion of notch ductility.

* *J. Iron Steel Inst.*, 1952, vol. 172, pp. 143–148.

I was surprised to read on p. 146 the statement:

"Treatment at 650° C. has removed all effects of strain. The high yield stress may be attributed to refinement of the ferrite."

In this instance, the author is dealing with tensile tests only, but the first part of the statement implies that this treatment at 650° C. removes all effects, including the notch-impact effects, which is certainly not true. The effect of cold work on the notch ductility of steel is a factor of great practical significance, especially in regard to mild-steel plates supplied to a specified level of notch ductility. The fabricator receiving this material, for which he has paid extra, may cold-bend the plates to shape without considering the consequences. I have recently collaborated with an engineering firm on this problem; it was found that a stress-relieving treatment (600–650° C.) did not restore the transition temperature of the strained steel to that of the unstrained material. This is confirmed by the work of Osborn and his co-workers.[1]

With the exception of these criticisms, the author is to be congratulated on her contribution to our knowledge of the effect of pre-strain and ageing on the shape of the stress/strain curve of specimens representing the length and cross directions of rolling, as shown in Fig. 2 of the paper.

Mr. **M. W. Thring** (B.I.S.R.A.): This paper begins to make a pier on the bridge that has to be built across the river separating metal physics from metallurgy and engineering, which is one of the most important fields of development at the present time. On the one side there is pure metal physics, *i.e.*, the dislocation theory and its various developments; this work is mainly concerned at present with very simple pure metals, *i.e.*, single crystals, mainly hexagonal or face-centred cubic. The work of Cottrell takes a step towards greater complication, being concerned with iron containing a very little carbon, although less than in the steel studied by Dr. Tipper. Cottrell, however, usually works with 0·02% of carbon, whereas Dr. Tipper's steel has 0·12%, which is an important difference.

The third stage is concerned with polycrystalline material, but again, however, with the dislocations anchored by small quantities of carbon and nitrogen, as in the Cottrell theory. The fourth stage is polycrystalline iron with more carbon and with a definite metallurgical structure, including pearlite and martensite as well as ferrite. This paper is half-way between the fourth stage and the fifth stage, where non-metallic dirt is put into the steel. Each of these five stages linking metal physics with metallurgy and engineering has its own appropriate theories.

The first stage, the pure metal physics, is concerned with the theory of dislocations, which has made remarkable strides in the last few years, the new conception of locking by cross-slip being very important in metal physics. It has so far only been studied with the face-centred cubic crystals, but it could also apply to body-centred cubic. This may possibly explain the author's results, since she mentions that if there are slip-lines at 45°, and also at 45° the other way, they should work equally for straining at right-angles. When those two sets of slip-lines begin to lock each other this does not apply, however, because the way in which they lock for strains in one direction will not be the same as the way in which they lock for strains in the other direction.

At the second stage, the Cottrell stage, we have already had the views of Mr. Cracknell and Dr. Petch. I think that the phenomenon with which the paper is concerned is not a Cottrell phenomenon, and that the yield point in these specimens is not due to the Cottrell mechanism at all; it is probably due to the higher quantities of carbon and to a metallurgical phenomenon. These

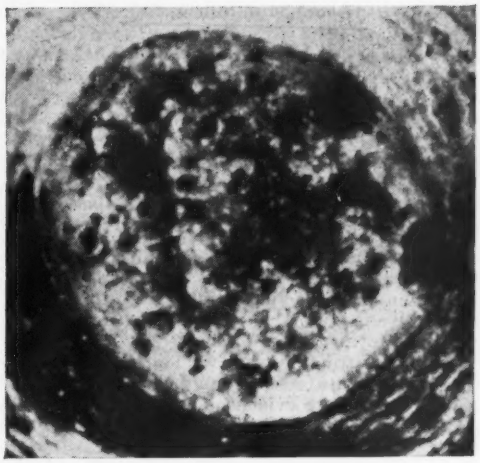

Fig. A—Tensile fracture in a copper specimen x 30

results might be explained by any one of three theories:

(i) The theory of the dislocations in the two planes locking each other, which is inappropriate for the pulling in the direction at right-angles

(ii) The theory set out in the written communication by Mr. Cracknell and Dr. Petch

(iii) A metallurgical effect due to the presence of carbon in larger quantities, which gives a metallurgical structure that is in some way rearranged by the straining.

I do not like the microcrack theory that Dr. Tipper mentions; I think that it should be a micro-strong-plate rather than a micro-weak-plate theory. It is an arrangement in the metal caused by the straining of some things that resist the movement of slip-lines, rather than a rearrangement of weaknesses. The structure of the metal that is being deformed should be examined by all means possible. The real value of this work lies in seeing these things in the metal and showing that they are there, doing what we anticipate.

I agree with Mr. Barr that the interesting curves in this paper are those showing the effects on the yield point of straining in the different directions. The raising of the yield point by straining in the same direction by 10% indicates that this is not a Cottrell mechanism, but that the sharp yield point in this metal is definitely due to another factor, which is probably the metallurgical structure, *i.e.*, the higher content of carbon. When it is pulled at right-angles it ceases to be sharp, which indicates that there is some metallurgical mechanism that locks the slip planes. When it is pulled farther in the same direction, they are locked very strongly, but there is still a sharp yield point. When pulled at right-angles they are locked more strongly but not all by the same amount, and there is no sharp yield point.

Dr. **B. B. Hundy** (B.I.S.R.A.): I was associated with the work at the Massachusetts Institute of Technology mentioned in the paper, and I should like to outline our ideas on the effect of microcracks on the mechanical anisotropy in metals. We think that a cast metal contains a number of randomly oriented microcracks. As these are not obvious under the microscope, they should perhaps be called sub-microcracks, but it is more convenient to refer to them as microcracks. We consider that the deformation of the cast metal will probably cause fresh microcracks to form, and all these cracks, regardless of their origin, will be aligned in the direction of metal flow during working. This crack structure can

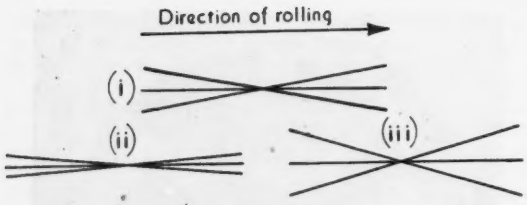

Fig. B—Distribution of microcracks : (i) before straining ; (ii) after pulling in rolling direction *L* ; (iii) after pulling in transverse direction *X*

be found in most ductile metals, and annealing does not seem to affect it. The fracture stress and the reduction in area to fracture were affected by the orientation of this crack structure, but we found that the yield strength or ultimate tensile strength were not appreciably affected. Most of our evidence for the existence of these cracks is indirect, but examination of the appearance of the fracture in a number of ductile metals tested in tension supported this theory.

Figure *A* shows the surface of a fractured tensile specimen in which a number of holes can be seen, apparently penetrating fairly deeply into the specimen. These specimens were pulled in the direction of forging. We consider that these holes are the result of the enlargement of microcracks by the radial and circumferential stresses in the neck.

Considering the author's results in the light of this theory, I should expect the hot-rolled plate to have an oriented crack structure in the rolling direction. Thus, if the microcracks affected the yield phenomenon at all, even in a minor manner, there should be a difference between the samples tested in the transverse and rolling directions. The results in Fig. 2, however, show no difference at all in the yield behaviour of the specimens tested in these two directions. Further pre-strain in the *X* and *L* directions will not greatly alter the distribution of microcracks, and so this crack structure should not influence the behaviour of these specimens during yielding. The yield phenomena described in this paper cannot therefore be explained on the basis of the microcrack theory. This theory, however, does supply a possible explanation for some of the Izod test results.

Figure *B* (i) shows the distribution of microcracks to be expected in this plate. If the specimen were strained in the rolling direction, these microcracks would close up a little, giving the effect shown in Fig. *B* (ii). If strained in the transverse direction, they should open up a little, giving the effect shown in Fig. *B* (iii). A crack structure of this kind might affect the Izod test results above the transition temperature. Figure *C* shows the distribution of the microcracks for the various Izod specimens according to this idea.

If there is a fault, the fracture crack will tend to propagate along it, and if all these faults are parallel to the direction of fracture propagation, the energy absorption in the impact test will be low. The specimen *Lx* shown in Fig. *C* (iv) would thus be expected to have a lower energy absorption than that in Fig. *C* (iii), and these two specimens, *Lx* and *Xl*, would be expected to have much lower energy absorption than either *Ll* or *Xx*, shown in Figs. *C* (i) and *C* (ii). If these ideas are developed, the author's results tie up nicely with this theory.

This does not explain the rise in the transition temperature on straining and ageing, but a study of some of the literature on the subject shows that this is found to some extent in most steels showing strain-ageing phenomena, and I imagine that it is in some way related to this. The microcrack theory may thus explain some of the observations in the Izod impact tests, but it cannot explain the difference in the effect of the pre-strain direction on the yield point in tension.

On p. 147, the author points out that there is a discrepancy of about 3 : 1 in the yield strength in tension and compression, but the figures given in Table II suggest that this is nearer 2 : 1. If this is so, the results might be explained on the basis of the Bauschinger effect. In some tests on pre-stretched aluminium alloy sheets, Welter[2] showed that the yield stress at a strain of 0·001% was 42,000 lb./sq. in. when tested in tension in the direction of pre-stressing, and only 32,000 lb./sq. in. when tested in the transverse direction; his specimen was stressed only a small amount. He also showed that if the specimens were pulled in the direction of pre-stressing to a greater strain, the yield strength on subsequent compression was only 23,000 lb./sq. in. These figures seem to be comparable with the differences given by Dr. Tipper. Researches carried out in America on copper and brass[3] showed similar effects, and even a stress-relieving treatment for 10 hr. at 250° C. did not eliminate the Bauschinger effect in copper. It seems possible, therefore, that the difference in the yield behaviour of Dr. Tipper's specimens may be explained by the Bauschinger effect.

CORRESPONDENCE

Mr. **A. Cracknell** and Dr. **N. J. Petch** (Leeds University) wrote: Dr. Tipper points out the remarkable effect of the direction of pre-strain on the subsequent yield point of mild steel. Thus, if the direction of pre-strain is at right-angles to the direction of subsequent straining (carried out after an intermediate ageing at 250° C.), the characteristic double yield point is absent.

We are at present carrying out some work on the yield point that may have a bearing on this observation. We started by considering the estimation of the room-temperature solubility of carbon from high-temperature

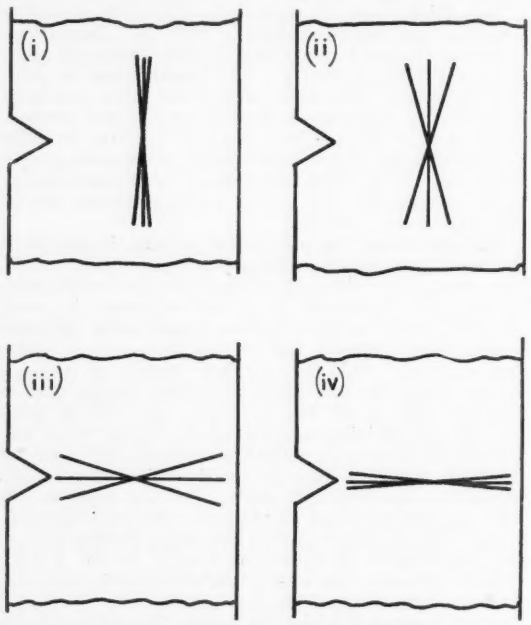

Fig. C—Distribution of microcracks in Izod specimens: (i) *Ll* ; (ii) *Xx* ; (iii) *Xl* ; (iv) *Lx*

data;[4] this indicates a figure that is too low to supply a carbon atom for each unit of dislocation. The actual solubility may be higher than this estimate,[5] but the possibility of incomplete occupation remains; indeed, a dislocation may be able to interact with more than one carbon atom. This suggests that in the Cottrell-Bilby theory, which associates the yield point with the stress required to free dislocations from their anchoring carbon atoms, the separation of a dislocation loop from an incompletely occupied dislocation line has to be considered. The greater the degree of occupation, the higher the yield point should be. In support of this idea, we have observed that quenching from 680° C. raises the lower yield point by about 10%.

A possible explanation of Dr. Tipper's observations may be on the following lines. There will be a multiplication of dislocations on the glide planes that are active during the pre-straining. Owing to the high stresses produced by the interaction between these dislocations, there may be carbon migration during ageing from the inactive to the active glide planes and to regions of stress concentration in the grain boundaries at the ends of these planes. If yielding takes place on the same families of slip planes as the original yielding, this migration should be reflected in subsequent loading by an increased yield point and, if there has been sufficient carbon migration, by the absence of a yield point when formerly inactive slip planes are brought into play by straining at right-angles to the original direction.

To obtain experimental evidence on this idea, we are examining the influence on Dr. Tipper's effect of increasing the supply of available solute atoms by nitriding.

Mr. **B. Watkins** (Stewarts and Lloyds Ltd.) wrote: From Table II in Dr. Tipper's paper, it would appear that the effect of ageing after pre-straining is to lower the yield stress in the directions designated *Lx* and *Xx*. The effect of ageing on the maximum stress is, however, of the same order in all the directions tested:

Direction	Stress for 10·5% Strain	Yield Stress after Ageing	Increase	Increase in Max. Stress
	Tons/sq. in.			
Ll	25·55	31·15	+5·6	+7·51
Lx	25·55	25·0	−0·55	+6·83

If the stress at 10·5% is corrected to the true stress at this point, the increase in yield stress on ageing becomes + 2·95 and − 3·2 tons/sq. in. in the directions *Ll* and *Lx*, respectively.

If, as appears from the shape of the curves, no ageing occurs at the yield stress (in the *x* direction), this lowering of the yield stress could be attributed to the Bauschinger effect. To verify this, and in an effort to separate the Bauschinger and ageing effects, the following experiments were carried out.

A 0·106% carbon O.H. killed steel was strained 9%, and tensile test pieces were cut in both the transverse and longitudinal directions. Care was taken to keep the specimens as cold as possible during machining. Tests were carried out within 2 hr. of straining, and after ageing for ½ hr. at 250° C.

The tensile properties were as follows:

	As Strained	As Strained and Aged
Longitudinal:		
Yield stress, tons/sq. in.	26·2	29·2*
Max. stress, ,,	27·6	31·5
Elongation, % ($4\sqrt{A}$)	26	23
Red. in area, %	69	61
Transverse:		
Yield stress, tons/sq.in.	24·5	24·3
Max. stress, ,,	27·4	30·5
Elongation, % ($4\sqrt{A}$)	25	23
Red. in area, %	64	61

*A marked yield point was observed in this condition.

In the as-strained condition, the maximum stress is the same in both directions, but the yield stress is greater in the longitudinal direction. This difference of 1·7 tons/sq. in. can be attributed to the Bauschinger effect. In the transverse direction, ageing has no effect on the yield stress, but the maximum stress is increased from 27·4 to 30·5 tons/sq. in. On ageing, a yield point was obtained in the longitudinal direction, both the yield stress and maximum stress being raised.

Similar results were obtained using O.H. and Bessemer rimming steels, but the Bauschinger effect was less than 1 ton/sq. in., probably because of the lower carbon content. Before testing, all the steels were in the normalized condition.

Our results confirm Dr. Tipper's findings and help to explain the large differences in yield stress that she observed in the directions designated *l* and *x* in her paper.

Mr. **N. H. Polakowski** (University College, Swansea) wrote: I was mainly interested in the results given in Fig. 2 and Table II, and in their interpretation by the author. I feel, however, that the discussion and summary given do not contribute towards explaining the nature of the observations made, and also that Muir's results have been misinterpreted.

On p. 147, when describing Muir's work, Dr. Tipper says, " If a compression specimen were cut . . . recovery took place until the yield in compression was equal to the raised yield in tension". Figure XIII in Muir's earlier paper[6] shows that this is not even approximately true. When a bar was strained to 33 tons/sq. in. by tension, allowed to ' recover,' and then retested in compression, it yielded at less than 20 tons/sq. in. Similar results are cited on p. 288 of his later paper,[7] and the present writer has in fact never succeeded in making the yield points of a pre-stretched rod equal in tension and compression by just allowing it to age.

The definition of the Bauschinger effect given by Dr. Tipper on p. 147 is of little use in connection with tests involving relatively large plastic strains and yield points, and not involving elastic limits. This definition is, moreover, neither strictly nor generally true; a more accurate one is given by Seitz.[8]

The suggestion that a jogged yield point, after being suppressed by the Bauschinger effect, can return only after a recrystallization treatment (*see* p. 147 of the paper) is not always right. This depends on a variety of factors, but, in many low-carbon steels stretched up to 8%, heating to only 350° C., or even less in some cases, brings back the discontinuous yield.

If a little care is exercised, the measuring of yield-point strain in compression is quite easy, and the disturbing effect of the ends is unduly exaggerated.

There was no reason to give prominence to the ' microcrack theory ' in connection with facts to which it bears no relation whatsoever. Years before Baldwin and Ripling began the tests to which Dr. Tipper refers at the end of her paper, Davidenkov and Shevandin[9] had conclusively disproved the microcrack concept in connection with some speculations of Kuntze. Their paper is an important one, and was abstracted in some detail in this *Journal*.[10]

Contrary to Dr. Tipper's firm statement that " . . . the Bauschinger effect . . . cannot explain the absence of a recovered yield after ageing," the writer has no difficulty in visualizing how it can do so. Quite briefly, plastic flow in one direction leaves residual micro-stresses in the metal between adjacent grains. These stresses may vary in intensity from nearly zero to values of the order of the yield stress. When the first straining is interrupted and the load is taken off and then put on again, all the

grains will recommence flowing under the same external stress as that applied at the end of the first straining. If the direction of deformation is reversed, however, or even only altered, the overall yield stress will be depressed because of the bulk Bauschinger effect, and different grains will begin to flow at widely different stresses because of the varying contribution of the residual micro-stresses from one grain to the next.

To my knowledge, there is no evidence to indicate that ageing a 0·12% carbon steel stretched about 10% at 250° C. for 30 min. should reduce these intergranular stresses to any sizeable degree. In effect, even if each grain ages fully and produces a small yield point, the resultant stress-strain envelope in reversed deformation will nevertheless be continuous and smooth, since each of the elementary yield points will occur at a different applied stress. Naturally, the whole as-aged curve will be displaced upwards in respect of the unaged Bauschinger curve. On the same grounds, continued unidirectional straining following an intermediate ageing treatment will produce a yield discontinuity. This argument can easily be modified to account for the absence of a recovered yield point when the direction of the second deformation is intermediate between the two extremities.

Experiments with non-ageing Al-killed steel will give quantitative but not qualitative differences compared with what is observed with ordinary steel. The suggestion of using " a metal which is isotropic, both mechanically and structurally " for similar tests is interesting. But what metals do remain isotropic after being cold-worked ?

AUTHOR'S REPLY

Dr. **Constance F. Tipper** replied: I should like to thank those who have taken part in the discussion for their contributions, because suggestions on how to solve some of these problems are needed. In regard to the contribution by Mr. Cracknell and Dr. Petch, I have said in the paper that the only explanation of the increased tendency to cleavage fracture at higher temperatures that I can offer is that it may be attributed to some precipitation of the carbide on certain planes.

One reason why I did not draw curves through my impact-test results was that I did not think that there were enough points. I do not entirely agree with Mr. Barr's interpretation, although I agree that, where the test samples are all fibrous, I should have expected very little difference, whether the material had been pre-strained or not. I do, however, think that the lower energy absorbed in the case of the partly crystalline fractures is due largely to the fact that the material is already half-way to fracture, and so much less energy is absorbed before the crystalline fracture begins.

Mr. Barr accuses me of making a misleading statement about the removal of the effects of pre-strain by heating at 650° C., and I agree with him entirely. The tensile-test results on the two samples, both as-rolled and after heating, were identical, and so the effects of the pre-strain in tension were removed. This does not mean that there were no other changes. When I have asserted previously that stress-relieving at 650° C. might even do damage (by lowering the impact figures), I have been criticized for making that statement.

I am very grateful to Mr. Thring for his remarks, and am glad to act as a bridge occasionally, even if in a limited way. Cottrell's work is admittedly done on steel with much lower carbon contents, but it is only the solubility of the carbon in the ferrite that matters. Although my steels have a higher carbon content, so far as we know the ferrite can take in only a certain number of carbon atoms in solution, so that there is not so much difference as Mr. Thring's remarks might suggest.

I agree that some of the effects that I have found here are metallurgical, and are not necessarily due to the Cottrell mechanism, but the Cottrell explanation of strain ageing and the recovery of the yield point does not, unless it is masked by some other effect, such as the Bauschinger effect, account for the change in the character of the yield point now observed.

I am very glad that the work at M.I.T. has been mentioned, because so far as I know it has not yet been published. When I saw Dr. Shaler in the spring, he mentioned that he had not published his work because he had found an anomaly in the case of aluminium, which did not behave like the other metals, and he said that he was still cogitating about the whole problem.

Holes in the fracture surface are very common, and I drew attention to the phenomenon some years ago in connection with the propagation of fractures in mild-steel plates[11]. I have since investigated fractures in copper plate starting from a notch in which the holes or cracks were developed well into the plate, at least 0·2–0·3 in. below the visible crack on the surface.

There is a difference of 10 tons between compression and tension in 30 tons, and I agree that a good deal of this can be accounted for by the Bauschinger effect. What I should like to be able to do now is to separate out the various factors.

Dr. **Tipper** wrote in further reply: I regret that the experiments did not lead to any definite conclusions about the relative weight to be attached to the possible causes of these observations. I made it quite clear that no single theory could explain the results, probably because several processes were going on simultaneously. Mr. Watkins's contribution was therefore especially welcome, because he had carried out just the type of experiment that was needed to sort out these processes.

I am very sorry if I have inadvertently misinterpreted Muir's pioneer work. Muir's results are also confused, because they are affected by both ageing and the direction of straining. I thank Mr. Polakowski for drawing my attention to the definition of the Bauschinger effect given by Seitz, but I do not think that it materially alters the general picture.

The Bauschinger effect can be invoked to explain the differences in values of the yields on reversing the direction of straining, but it cannot, in my opinion, account for the difference in shape. The microcrack theory is very attractive, but it does not account for the higher ultimate breaking stresses in tension.

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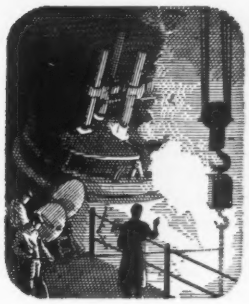

Industrial Electric Furnaces

By J. C. Howard, B.Met.

SYNOPSIS

The paper reviews the use of electric energy for metallurgical heating purposes, with particular reference to the more recent developments in this field, and gives some indication of possible future trends in development and application. Arc, resistance, and induction furnaces are all discussed. 741

ELECTRIC heating for metallurgical purposes was little used on a commercial basis before the first World War, although at that time arc furnaces of various types had already been installed for steel melting. The war gave a great fillip to the new process and many furnaces were installed, a few of which are still working.

At about this time it was also found that electric resistance furnaces enabled heat-treatment processes to be carried out with an accuracy of control that had not hitherto been possible.

Later came the commercial application of induction heating. The earliest was in the form of normal-frequency furnaces of the ring type for metal melting. Then came high-frequency furnaces for metal melting, and later still the induction principle was used for heat treatment and for heating before hot forming.

There is a wide field for future development, but it is unlikely that there will be any new principle of heating used. There are three basic principles in use: arc heating, resistance heating, and induction heating. Resistance heating can be indirect, using heating elements, or direct, using the resistance of the charge. The latter is similar to induction heating, in which induced electric currents flow in the charge or work-piece and so heat up the material by its own resistance to the flow of these currents.

RESISTANCE FURNACES

The resistance furnace, in which the heat is generated within the resistor element and transferred to the charge by radiation and/or convection, is now an important item of heat-treatment equipment.

In the early days of its development, automatic temperature control was little known outside the laboratory. Temperatures were largely judged by eye, sometimes with the help of an indicating pyrometer, and control was effected by manual adjustment of the burners, or, with solid-fuel furnaces, by varying the rate of firing. There was little mechanization, and scaling and decarburization were accepted as inevitable in a process in which the steel had to be in continuous contact with the products of combustion.

The cleaner operating conditions of the electric furnace, together with the more uniform distribution of heat and the absence of combustion products within the furnace chamber, gradually came to be recognized as advantages, although they were not always considered to be sufficient to offset the much lower price of available fuel. It was not until the electric-furnace industry had pioneered automatic temperature control that the resistance furnace became accepted, and at this stage the other advantages of electric heating began to be more fully exploited. Perhaps the most important of these advantages was the fact that a protective atmosphere could be used to prevent scaling without an inner muffle being incorporated in the furnace chamber.

The recent development of the gas-fired radiant tube has, however, permitted the use of controlled atmospheres in gas-fired furnaces, so that this advantage is no longer exclusive to the electric furnace.

The many recent developments in furnace technology have been fully applied to electric-furnace practice. Temperature-control equipment has been improved, with such refinements as programme control for special-purpose treatments. The sub-division of the larger furnaces into two or more independently controlled zones, each with separate thermocouples

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and control equipment, enables the heat input to be matched to the varying demands in the different parts of the furnace and any required temperature distribution to be obtained.

Excess-temperature devices are now always included in electric furnaces. These may take the form of a simple heat fuse arranged to trip out the furnace contactor should excess temperature be reached, or to operate a light or other warning device through a relay. The temperature instrument may incorporate an excess-temperature contact that opens on rising temperature at a selected point above the control temperature and trips out the furnace contactor when such a temperature is reached.

The development of heat-resisting steels and alloys has made increased mechanization possible, and the current tendency in furnace design is to eliminate the human element as far as possible. Much modern equipment is built to give maximum efficiency in performing one specific heat-treatment operation on a particular type of component; this is especially true of continuous furnaces.

Controlled Atmospheres

There has recently been much progress in the application of controlled atmospheres to steel processing. Atmospheres derived from raw town gas can be generated outside the furnace, at less than half the cost of the same volume of raw gas, for the purpose of preserving surface appearance during heat treatment.

More recent developments in the design of controlled-atmosphere plant enable a slightly more expensive atmosphere to be generated outside the furnace for heat-treating high-carbon and alloy steels, with complete freedom from decarburization, irrespective of the initial carbon content and the treatment temperatures involved. For those alloy steels that cannot readily be bright-annealed with atmospheres derived from town gas, the atmospheres derived from anhydrous ammonia are available, so that it is now possible to undertake the clean heat treatment of most ferrous alloys, provided that the appropriate atmosphere is used.

Similar progress has been made in the generation of atmospheres for gas carburizing, carbonitriding, and straight nitriding, giving reduced production costs, superior products, and the elimination of such operations as pickling.

Although the standard design of electric resistance furnace requires little modification to make it suitable for bright heat treatment, the introduction of these processes on a large scale has called for the development of special plant, as in the case of gas carburizing and carbonitriding.

Salt Bath Practice

Parallel with progress in dry furnace treatment has been the progress in the application of electricity to salt-bath practice; a faster rate of heating is possible in the closely spaced electrode-type salt bath. Advances in salt-bath practice have called for the development of new salts and the improvement of the older salts to get the quickest possible rates in such processes as cyaniding, liquid carburizing, and the superficial nitriding of high-speed steel. The increasing practice of salt-bath-hardening carbon and alloy steels and high-speed steels has necessitated the improvement of neutral salts so that strict neutrality of the salt can be maintained under operating conditions.

The descaling of stainless and other steels in molten salt is another recent development in this field. The work of Davenport and Bain, which has led to the wide use of isothermal heat treatment in the U.S.A., has not been adopted so widely in Great Britain. One or two plants have, however, been installed for the purpose of martempering instead of the conventional oil-hardening and tempering; they give pronounced reduction and, in some cases, virtual elimination of the distortion problem.

Methods of Heating

In regard to methods of heating, resistors are usually of the 80/20 nickel–chromium type, which is satisfactory for temperatures up to 1100–1150° C., although the iron–nickel–chromium type is often used at lower temperatures. The elements, which may be in the form of sinuous strip or rod, coiled rod or wire, or occasionally tube, may be mounted in the roof and side walls in slots in the brickwork, or suspended by means of pins or hooks. They may also be placed in the door and hearth should operating conditions and type of furnace render this necessary.

For higher temperatures, the iron–chromium–aluminium resistor, which has an upper limit about 200° C. higher, is often used, but, because of its reduced strength, this requires careful handling. Non-metallic resistors of silicon-carbide rod are therefore sometimes preferred, and, for temperatures in the high-speed steel range, these are fairly standard in small and medium-sized furnaces. They are, however, subject to wear during use, with a consequent increase in electrical resistance, and so, to ensure that the full rating is maintained, some form of voltage regulation, such as a multi-tapped transformer, is essential. One important advantage of this type of resistor is the speed and ease with which it can be replaced.

With operating temperatures below 750° C., radiation plays a diminishing part, and heat is largely transferred by convection. Developments designed to speed up heat transfer while maintaining a high degree of temperature uniformity are seen in improved air-circulating systems and the restriction of the transmission of radiant heat. High-speed fans are now standard, and a recent development has been to isolate the work chamber from the heating unit, so that direct radiation from the resistors is eliminated.

The temperature limitation in resistor materials for the large furnaces, such as those used for reheating before forging, etc., has restricted the application of electric heating in this field, so that the recent introduction of the slag resistor furnace is interesting. This furnace, designed for operating temperatures between 1000° and 1500° C., uses a bath of molten glass as the source of radiant heat. The glass is contained in carbon troughs on either side of the hearth; it is heated by current flowing between immersed graphite electrodes. The lining of the furnace rapidly becomes glazed and heat is radiated directly from the bath surface and indirectly from the roof

and side walls to the hearth, resulting in a high energy dissipation. To cope with variations in the electrical resistance of the molten glass with temperature, the supply transformer is provided with a wide range of tappings.

Batch Furnaces

For other heat-treatment processes, furnaces may be conveniently considered in two broad classifications, *i.e.*, the 'in-and-out' or batch type, and the continuous type; both are used with or without controlled atmospheres.

For annealing, normalizing, and hardening there are three common types of open batch furnace:

(i) The bogie hearth furnace, in which the hearth is withdrawn for charging and discharging by means of an overhead crane
(ii) The horizontal batch or muffle type, generally used in conjunction with a vertical-lift charging machine
(iii) The pit furnace, used when long slender parts are to be heat-treated in a vertical position to avoid distortion.

The bogie hearth has the disadvantage that the hearth is generally unheated, but it is nevertheless suitable for the heat treatment of heavy or awkward shapes. The horizontal batch type is usually preferred for normalizing and hardening general work, such as forgings or castings in bulk.

In all three types, some form of protective atmosphere may be used to reduce the incidence of scaling, but where truly clean heat treatment is required, specially designed furnaces are preferable.

For clean hardening, a modification to the horizontal batch furnace includes a tilting hearth with an incorporated quench chute to allow the quench to be made out of contact with the air.

For bright annealing, particularly of coiled products, there is the bell-base or lift-off type of furnace and the pit-pot furnace. The bell-base furnace consists of a movable bell-shaped furnace with automatic temperature control and a number of fixed bases on which the charge is stacked. Each base is provided with a fan, if the operating temperature warrants it; a bell-shaped hood or cover which fits over the charge is provided at the bottom with a flange to fit into an annular seal surrounding the base. The number of bases provided is in the ratio of the cooling to the heating cycles and is sufficient to permit continuity of production.

In operation, a base is loaded, the hood is placed in position, connection is made to the externally-generated prepared atmosphere, and the container is purged of air. The bell, already at temperature, is then lowered over the hood and its temperature is maintained for the heating and soaking cycle required. The bell is then transferred to a second base already loaded and purged, and the atmosphere outlet from the first base is closed, allowing the charge to cool under a positive pressure of prepared atmosphere.

In the pit-pot furnace, the furnace body is usually sunk below ground level with the top open. The base containing the charge is suspended from an insulated pot head, which is lowered into a light-weight container. This container is then sealed, purged, and loaded into the furnace. The sequence of operations is the same as for the bell-base furnace, and the number of pots is sufficient to allow continuity of production.

Continuous Furnaces

There are many types of continuous furnace, and they are usually designated according to the method adopted for conveying the work through the furnace. Little modification in construction is required to render some of these suitable for use with protective atmospheres, but in others the provision of a cooling chamber, usually water-cooled, or an incorporated quench following the heating chamber, is essential to preserve surface finish.

Various types of belt-conveyor furnace are in use for the heat treatment of parts in bulk; mesh belts, chain belts, and slat conveyors are standard types, the choice being dependent upon the size of the furnace, the temperature of treatment, and the load of work per unit area of belt.

The pusher furnace, in which the charge, contained in trays, is pushed by mechanical or hydraulic pusher gear over idle rollers or skid rails along the hearth, is used for hardening, tempering, or normalizing small forgings. The driven-roller hearth has a variety of applications, such as the bright annealing or normalizing of tubes, bars, strip, etc.

For the continuous single-strand heat treatment of strip or wire, the 'pull-through' furnace is used. In this, the material forms its own conveyor, being uncoiled at the charging end and re-coiled after discharging. This type of furnace is particularly suitable when quick cooling is required, as in the annealing of austenitic stainless wire or the patenting of high-carbon steel wire, or for quenching, as in the hardening and tempering of spring-steel strip.

For the clean hardening of small parts, the reciprocating or shaker hearth is used. The parts are conveyed through the heating chamber by the reciprocating motion imparted to the hearth plate by pneumatic or mechanical means; they are eventually quenched *via* the discharge chute at the end of the hearth plate without losing contact with the protective atmosphere.

Case Hardening

The use of gas carburizing and carbonitriding has been greatly extended for surface hardening, giving lower production costs and improved quality of product. The carburizing atmosphere is usually derived from raw town gas, using an external generator, with or without the addition of an enriching gas, such as propane or butane. Alternatively, the atmosphere may be obtained by the thermal decomposition of alcohol mixtures dripped directly into the furnace.

The most common batch-type furnace is the pit type. This uses a fixed gas-tight retort, and the charge is contained in removable charge carriers, which are loaded in the hot furnace before the generated atmosphere is fed in. A circulating fan ensures the uniform distribution of the atmosphere. This arrangement permits direct quenching after carburizing. An alternative arrangement is to use a movable retort, which is loaded and purged in the cold before charging.

The carbonitriding process is similar to the gas carburizing process, but an addition of ammonia to

the atmosphere permits case hardening to be conducted at a lower temperature. Nitrogen as well as carbon is absorbed into the case, giving a more wear-resistant surface. Other important properties of carbonitriding are the reduction of distortion, due to the fact that the operating temperature is about 50° C. lower than in gas carburizing, and the fact that the presence of nitrogen in the case permits the substitution of oil-quenching for water-quenching without any loss in surface hardness. There is an appreciable saving of cost compared with the very similar salt cyaniding technique. There are, moreover, indications that many parts at present carburized could be carbonitrided with benefit, giving consequent improvement in service.

In tool hardening, particularly in the high-speed steel range, the use of salt baths is increasing; it is now usual to install the complete plant, using immersed electrode heating for the pre-heating, high heating, quenching, and secondary hardening. This type of equipment is of maximum benefit when continuity of production is required.

The alternative type of furnace suitable for hardening small high-speed steel tools is the small muffle furnace, heated by silicon-carbide resistors. These furnaces contain a specially prepared carbon block or muffle which, at high-speed steel temperatures, yields a CO/CO_2 atmosphere that is non-decarburizing to high-speed steel. This type of furnace is particularly applicable in small toolrooms, where the throughput is not sufficient to merit complete salt-bath installations.

ARC FURNACES

The type of furnace given the most serious consideration today, especially for steel melting, is the furnace patented by Heroult, in which three electrodes arc on to the charge or bath of molten metal, using 3-phase current. Almost all the heat is generated in the arc, although there is some heating of the charge by resistance.

Most modern furnaces are round, with a dished bottom and a domed refractory roof through which the three electrodes pass. The electrodes are supported in holders, attached to carriages that move up and down guides on the transformer side of the furnace. Two-way tilting is provided, so that furnaces can be tilted in one direction for pouring out the finished melt through the spout into a ladle, and in the opposite direction for slagging off.

Most of the new furnaces over 1 ton in capacity are of the top-charged type, in which the electrodes and their mechanism, together with the roof, can be moved aside for the whole charge to be dumped into the furnace. There are three main types of top-charged furnace available, each of which has its advantages. These are the swing-roof type, the gantry-type removable roof, and the gantry-type removable body. Of these, the swing-roof type is perhaps the most popular, and is certainly the cheapest to build and install.

Although the oscillating hearth furnace, which was introduced about ten years ago, has certain advantages, it is mechanically complicated, for which reason its use has declined recently.

A Swedish innovation is induction stirring of the molten bath by means of suitable inductors fitted to the base of the furnace, which must have a non-magnetic bottom between the inductor and the bath. This method shows greatest promise with large furnaces and for special steels, where very low carbon or low sulphur and phosphorus are essential. It has a distinct advantage at the slagging-off stage, in that the slag is pushed towards the charging door, where it can easily be raked off.

Furnaces with capacities of 100 tons and greater are now quite practical, and their use is simplified by such devices as remote control of electrode holders, doors that are more easily controlled (power-operated on larger units), water cooling of the more vulnerable parts, such as electrode holders, coolers, roof ring, doors and door arches, etc. All arc furnaces are now fitted with automatic regulators, whose design is greatly improved. All regulators are now operated by both the current and voltage components in each electrode. There is a preference for the rotary regulator, which has a separate D.C. generator feeding the D.C. operating motor on each electrode. The generator field is adjusted automatically in relation to the current and voltage components of the particular electrode, so that the current to the electrode-raising motor is regulated to correct the electrode position quickly. An advantage of this system is that there are no relays or contactors that require maintenance. There is, however, considerable support for the hydraulic regulator, which has the advantage of giving an instant response to signals from the current and voltage relays because of the energy stored in the hydraulic accumulator. In practice, and especially with a plant that has been in use for some time, leakage of hydraulic fluid is difficult to avoid, and the plant is difficult to keep clean. The contactor-type regulator is now considered to be obsolete. There will probably be an increase in the use of electronic regulation in the future.

In regard to electrical equipment, the power/weight ratio of furnaces has increased, especially in foundries. Furthermore, as larger furnaces are built, larger transformers are required, which need higher secondary voltages if the power factor is to be kept reasonable. A voltage higher than necessary should not, however, be used, because this gives too long an arc, which causes undue fluxing of roof and side-wall refractories. Larger transformers have a wider range of arc voltages to suit the varying metallurgical stages, and with the larger units the extra expense of on-load tap-change gear can be justified. This gives an appreciable saving of time and consequently increases output.

Oil circuit-breakers are still the most widely used type, but, where the extra expense can be justified, and especially on the higher supply voltages, the air-blast breaker is preferred.

There have been considerable developments in refractories for the arc furnace. The newer materials are not limited to use in this one type of furnace; in many cases, these improved refractories have been developed for arc-furnace use, and have then been applied in other fields. Most arc furnaces are basic-lined, the dry graded-dolomite hearth of the Crespi or Swift type being widely used. Side walls are

generally made from basic materials, such as magnesite or chrome magnesite, and steel-cased or steel-tube bricks have greatly increased refractory life.

In the future, arc furnaces of the present sizes, improved in detail, will still be used widely, mainly in foundries and for quality-steel ingot production. There will be, however, an increase in the number of large furnaces. Furnaces larger than those now in use will be developed, since the limit of power input has not yet been reached, although further increase in power must also mean a higher arc voltage during the full-power melting stage. These larger furnaces will be able to produce the cheaper steels now made in the basic open-hearth furnace, which are required in enormous quantities. An electric furnace shop designed to produce, say, 5000–10,000 tons of ingots per month would cost less than an open-hearth shop designed to produce a similar tonnage, if buildings, cranes, foundations, etc., are taken into consideration as well as furnaces. The arc furnace can deal with a wider variety of scrap than the open-hearth; it can melt all-scrap charges and, since the cost of the charge is less, the final ingot cost is less. For real economy, the furnace must be a large one; better results would be obtained by using a group of furnaces, which would enable the cycle time with the various furnaces to be staggered. This would give an overall high-load factor for the whole works, keep down the maximum kVA. demand charge per ton of steel produced, and so keep power costs low.

A satisfactory technique has not yet been developed for converting pig-iron into steel, but this should be possible with the extended use of the arc furnace with the lower grades of steel.

The electric smelting furnace has been little used in Great Britain, mainly because of the high power consumption involved, which makes the process uneconomic with British power tariffs. Although some ferro-alloy smelting furnaces were installed during the war because of the emergency conditions, there is not likely to be any wide use of smelting furnaces in this country until power is much cheaper.

INDUCTION FURNACES

The principle of induction heating is well-established in metallurgical practice. Induction furnaces can be divided into two main types:

(i) Those using normal-frequency A.C. from the ordinary supply
(ii) High-frequency furnaces for which some form of H.F. converter is used.

There is also a broad division of induction heating into melting furnaces and heating for forming and heat treatment.

Melting Furnaces

The normal-frequency furnaces, as used in melting, are generally of the core type. They have a true transformer with an iron core, and the metal to be melted is heated in a closed loop, forming the secondary. The first furnaces built were for steel melting and had an open horizontal slot. These furnaces rapidly fell into disfavour because the very severe erosion of the slot lining made the refractory life short.

After the abortive start with steel, the use of this type of furnace was practically limited to brass melting, in which field it has been completely successful; about 90% of the total production of brass in the world is now melted in this way. Refractory difficulties were not encountered because of the low temperature of the molten brass, and because the fluxing action on the refractory when melting brass was not severe. With the improvement of refractories, the application of this type of furnace has become much wider. Nickel silver, cupro-nickel, pure copper, bronze, and even cast iron can all now be tackled successfully. An extensive increase in the use of normal-frequency furnaces for melting special cast irons is likely. They will also be used in conjunction with cupolas, where high temperature or low sulphur content is essential, especially for making some of the special and small castings for the motor-car industry. It is, however, unlikely that a refractory will be found that will enable this furnace to be widely used for steel melting.

Core-type normal-frequency induction heating is sometimes used for heat treatment and heating for forging, but opportunities are not very numerous because the work-piece must itself be ring-shaped to take full advantage of this method of heating. There are, however, a number of cases that are suitable, such as the heating of railway tyres for shrinking on to hubs and the heat treatment of starter-ring gears on motor cars. Core-type normal-frequency furnaces, both for melting and heat treatment, have in their favour cheapness and efficiency, and for these reasons new uses are likely to be found for them. Compared with the so-called 'core-less' or 'high-frequency' furnaces, their power factor is good, and for this reason they are cheaper, since power-factor rectifying condensers are unnecessary. The heat is generated in the charge to be heated, which is the most efficient method of conversion. This type of furnace is more efficient than the H.F. furnace because there is no loss caused by the H.F. converter.

The H.F. furnace is generally core-less, although in some of the larger units there are types of iron cores. The secondary metallic loop of the core-type normal-frequency furnace is dispensed with, and the charge, or rather the molten bath, is cylindrical. The complete charge can be poured when ready without leaving a 'heel' of metal for starting the next charge, which is essential with the core-type normal-frequency furnace.

The refractory lining is very simple and easy to replace. With small furnaces, the lining can be renewed in about an hour, and with large furnaces it can be replaced in about a shift. The lining life depends very much on the metal being melted, the temperature of pouring, etc., and may vary between 30 and 300 heats. When considering furnace refractories, three essentials must be borne in mind:

(i) The cost of the materials
(ii) The work involved in renewing the refractory
(iii) The actual lining life.

All these must be considered in combination to obtain a true picture of the problem. Similar materials costing about the same are used in both the core-type and the H.F. furnace. The former requires much more refractory than the H.F. furnace of equivalent

hourly output, and so the total refractories cost is much higher. The H.F. furnace lining can be replaced quickly, but with a normal-frequency furnace, replacement takes days, or even weeks. To be satisfactory, the lining life of the normal-frequency furnace, as measured in the number of heats, must be much longer than that of the H.F. furnace. For steel melting it would be practical to produce a refractory lining for the normal core-type induction furnace that would give a lining life of about 50 heats, but this would be unsatisfactory because of the relatively high cost of refractory materials and labour for renewing this lining.

High-frequency furnaces vary in size from a few ounces up to 10–12 tons. In ferrous metallurgy, the smaller sizes are used for research work and making small heats for precision casting. The very small units usually have a spark-gap converter or a valve oscillator as a source of H.F. power. Medium-sized furnaces (*i.e.*, 1–20 cwt. capacity) are used for special steels for foundry work and for making tool steels. The H.F. furnace has almost entirely replaced the old crucible process for making tool steel, because of the conversion cost and the absence of sulphur.

The larger H.F. furnaces are generally used for making the carbon and low-alloy steels used in the motor-car industry. The size of the furnace depends on the demand, but the larger H.F. furnaces can compete with the arc furnace. The two types are in fact complementary, because in any large steelworks making quality steels, the combination of arc and H.F. furnaces is particularly useful. The arc furnace should be used for making high-quality steels from low-grade materials by means of the refining processes that are possible with it, and the H.F. furnace should be used for melting the rolling mill and ingot scrap of the arc-furnace product.

The high-frequency furnace is best used as a remelting furnace, melting high-quality materials to make high-quality steels. Although it is possible to carry out metallurgical refining operations in it, this is not usually recommended. The refractory lining in the H.F. furnace is relatively thin, so that it is not suitable for withstanding the corrosive action of the slags that would have to be used.

Although spark-gap or valve oscillators are often used for the smaller units, the motor-generator rotary converter is usually preferred for sizes over 20–30 lb. The frequency obtainable from spark-gap or valve oscillators is generally about 40,000–400,000 cycles/sec., and with motor generators the frequency ranges from 10,000 cycles/sec. for the small furnaces to, say, 500 cycles/sec. for the large furnaces. More recently, static H.F. converters using mercury-arc 'inverters' have become available. They do not yet constitute a genuine alternative to motor generators, but future development may bring this about.

The earliest production furnaces had their inductor coils mounted in a wooden box, a simple construction that was very effective. The simplicity has been maintained for the newer designs. In furnaces up to 1 ton capacity, it is usual to have the furnace bodies made from asbestos cement boarding, mounted on a metallic framework, the joints of the framework being insulated to prevent circulating currents from heating it up. This simple construction gives cheapness of manufacture and ease of maintenance. For the larger furnaces, *i.e.*, those of 1 ton capacity and over, the simple board construction is not robust enough to withstand the heavy strains that may be imposed on it. These larger furnaces are usually made from steel platework, the body being cylindrical. Some form of screening between the coil and the steel body is essential to prevent the latter from being heated, and this is often done by interposing packets of transformer iron between the body and the coil to provide a return path for the magnetic flux. These transformer-iron cores serve a dual purpose, since they also provide support for the coil.

Furnace layout requires careful consideration. It is preferable to have the condensers (or capacitors, as they are now generally known) as close as possible to the furnace. They can be situated in a cellar behind the furnace or, should the furnace be up against the side walls of the main building, they can be in a lean-to building outside. The control panel must be close to the furnace and within easy access of the furnace operator. If normal precautions are taken, there is no danger to the operator, but it is usual to surround the furnace with an insulated wooden platform to give added protection. Furnaces are usually nose tilting, *i.e.*, they pivot about an axis through the spout, so that the position of the stream does not change with the angle of tilting. The top of the furnace is level with the top of the working stage. From the operator's point of view, the furnace can be looked upon as a hole in the platform at his feet, into which the charge material has to be dumped and in which it is also easy to manipulate the tools that are necessary for poking, etc. The surrounding timber platform is usually supported on a steel framework, and underneath the platform is the tilting equipment adjacent to the furnace. Electromechanical methods of tilting are generally used, but hydraulic tilting is often favoured, since it gives a very smooth control action.

It is not essential to have the motor generator near the furnace. It can be housed anywhere in a convenient building at a distance from the furnace, where conditions are cleaner than in the foundry and so more suitable for the machine. The development of totally enclosed machines with water-cooled heat exchangers has created dust-free conditions and has reduced the cost of installation. Increased use of this heat-exchanger type of machine is likely in the future.

The largest furnace of which the author knows is of 12 tons capacity with a 3500-kW. generator. It is doubtful whether furnaces will be very much larger in the future. Unfortunately, the H.F. furnace has a low power factor (less than 0·1 with large furnaces), so that a 3500-kW. installation has more than 35,000 kVA. capacity. Great ingenuity is needed to cram 35,000 kVA. into the space occupied by a 12-ton furnace body. The lineal dimensions of a 50-ton furnace would not increase proportionately. The problems involved in getting five times as much power into bigger furnaces would be very great indeed.

If larger furnaces prove to be practical, it is unlikely that the depth of the bath will be greatly increased. From a general design point of view it is preferable

to have the depth of the molten charge about equal to its diameter, but it has been found in practice that if the depth of molten metal is increased beyond about 40 in., severe refractory problems at the base of the furnace are encountered, owing to the pressure there. Any future extra capacity will be obtained by increasing the diameter.

Induction Heating

Induction heating of steel, both for heat treatment and hot working, is a much more recent application of this principle of heating. Both high- and normal-frequency, and sometimes a combination of both, are used. It is not, strictly, a furnace application; in many cases there is no furnace in the accepted sense. Apart from the work-piece, nothing is hot, and cold water is much more evident than heat. Induction heating, however, carries out a metallurgical function that would otherwise be done in a conventional furnace, so that reference to it is logical.

Induction surface-hardening of steel is now well known. It is, perhaps, more extensively applied in the U.S.A. than in Great Britain, probably because of the bigger American markets, since induction hardening can be applied with the greatest advantage to mass production. A suitable inductor placed near the work-piece induces electric currents in its skin, and the power concentration is so proportioned that the surface is heated to hardening temperature before the core is hot. Thus, on quenching, only the case is hardened, the core retaining its original metallurgical structure. The process has made possible a much wider use of plain-carbon steels, because it is possible, by suitable selection and prior treatment, to have (with the cheaper plain-carbon steels) case and core properties that can otherwise be obtained only with alloy steels. By this means, parts in the soft state can be machine-finished and subsequently hardened without distortion or oxidation.

Similar surface-hardening results are obtained by flame hardening. Flame hardening is better where the work is less mass-produced, but the hardened case is usually deeper, owing to the longer heating time required. The operational costs of flame hardening are higher.

The products of the motor industry are typical examples where induction hardening finds great use, and a wide extension of the application to this class of work is expected. There will probably be an increase in the size and number of installations in which the hardening is carried out as a stage in the production line; this avoids sending off the parts to a heat-treatment department.

Induction heating is also applied to through-heating of metallic parts where through-hardening and tempering is required. A typical case is the heat-treatment of steel bars, which are fed progressively, first through the inductor to be heated for hardening and then through the quench. The bar is then passed forward to another inductor, where it is heated to a lower temperature for tempering. This type of through heat treatment is usually carried out progressively and is most suitable for bulk production. The work-piece may be strip or sheet, using a special type of inductor or the more normal helical inductor. Perhaps the main advantage is the comparatively rapid rate of heating. The piece may be round or any symmetrical shape, and it may be solid or tubular in form.

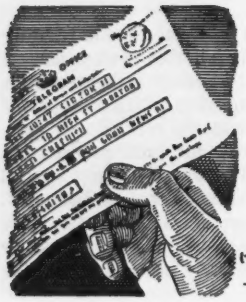
Induction heating is less suitable for heating large cross-sections of, say, 10 in. dia., because the heat is generated in the surface layer and the centre is heated only by thermal conduction from the hotter outside layer. Lower frequencies increase the diameter that can be efficiently heated but beyond a certain size the rate of heating is so governed by the rate of thermal conduction that the induction principle has no advantage over conventional furnace treatment.

Depending on the application, the frequency varies from 500,000 to 50 cycles/sec. Very high frequencies and small powers are used for heating small sections, and these are obtained from valve oscillators. The medium frequencies are obtained from motor generators, and, where 50 cycles is used, this is obtained from the normal power supply. It is in some ways curious that normal-frequency heating should be the last to be applied to any appreciable extent. It will no doubt be used more widely in the future, because the initial plant cost and the operating costs are less. Development should overcome some of the fundamental difficulties of application that now exist.

Induction heating for hot working is now well established, particularly in the field of forging, where, once again, it is most suited for mass production. Any forging operation in which dies are used is likely to be suitable for induction heating. The advantages are:

(i) The reduction of scale formation, which reduces the loss of steel and increases the die life
(ii) A relatively small number of billets are tied up in the heating process, so that starting and stopping can be done quickly, without loss of time
(iii) The operating conditions are greatly superior.

Induction heating has been considered for use in rolling mills. It is suitable for the smaller mills, such as bar mills, but the tonnage output of even these small mills is so high that a very large and expensive installation would be necessary. So far, capital cost has prevented its use in this field, but normal-frequency, or a combination of normal- and high-frequency, heating should help to surmount these difficulties.



NEWS

ANNOUNCEMENTS AND NEWS OF SCIENCE AND INDUSTRY

THE IRON AND STEEL INSTITUTE

Meetings

Meeting on Boron

It is regretted that this meeting, which was to have been held on February 18th, has had to be postponed.

Annual General Meeting, 1953

The **Annual General Meeting** of The Iron and Steel Institute will be held at 4 Grosvenor Gardens, London, S.W.1, on **Thursday 30th April and Friday 1st May, 1953.** Details of the programme will be given in the April issue of the *Journal*.

Andrew Carnegie Research Fund

Mr. S. GARBER (Sheffield University) has been awarded a second grant from the Fund for his research on the formation of graphite in cast irons.

Mr. A. SHELTON (City and Guilds College, London) has been awarded a second grant for research on the measurement of Poisson's ratio in the elastic range of various metallic materials and its extension into the elastic–plastic range and fully plastic range at various degrees of cold-working.

NEWS OF MEMBERS

▶ Mr. R. BARTRAM has been appointed Chief Engineer of the Material Handling Department of Fraser and Chalmers Engineering Works, Erith.

▶ Professor P. G. BASTIEN has been awarded the V. Noury Prize by the Institut de France (Academie des Sciences) for his contribution to metallographic research.

▶ Mr. C. R. BLICK has been appointed Assistant Commercial Manager of the Workington Iron and Steel Branch of the United Steel Companies, Ltd.

▶ Mr. C. BODSWORTH has been appointed Lecturer in Metallurgy at the University of Liverpool.

▶ Dr. L. D. C. BOK has left the University of Natal, Durban, and is now at the Departement Skeikunde, Universiteit van die Oranje-Vrystaat, Bloemfontein.

▶ Mr. P. BROCK has been awarded the degree of Ph.D. of the University of Sheffield. He has taken up an appointment with the British Non-Ferrous Metals Research Association.

▶ Mr. DOUGLAS BRUCE-GARDNER has been appointed a Director of John Lysaght's Scunthorpe Works Ltd.

▶ Dr. R. D. CARTER has left the Plessey Co., Ltd., Ilford, to take up an appointment in the Research Organization of British Insulated Callenders Cables Ltd., Shepherds Bush.

▶ Mr. D. K. COUTTS, formerly of the Indian Iron and Steel Co., Ltd., has taken up the appointment of Assistant Manager of the Technical Office of The Mond Nickel Co., Ltd., in Bombay.

▶ Mr. I. S. EDGAR is now designer to The British Tools and Pressings Co., Gerrards Cross.

▶ Professor E. HOUDREMONT has been awarded the honorary degree of Dr.-Ingenieur by the Technical University of Berlin-Charlottenburg.

▶ Dr. D. E. R. HUGHES has left Tube Investments Ltd. to take up an appointment with the G.K.N. Group Research Laboratory, Wolverhampton.

▶ Mr. W. L. JAMES, Director and General Manager of John Lysaght's Scunthorpe Works Ltd., has been appointed Managing Director of the Company.

▶ Engineer Rear-Admiral C. W. LAMBERT is retiring from the post of General Works Manager at the Appleby-Frodingham Branch of the United Steel Companies, Ltd., at his own request, on 30th June, 1953. He will be succeeded by Mr. A. JACKSON, at present Works Manager (Steel) and Deputy General Works Manager.

▶ Mr. BERNARD MAXWELL has completed his studies at the Royal Technical College, Glasgow, and has joined the Clyde Iron Works of Colvilles Ltd.

▶ Mr. S. G. MEGGITT has been appointed a Director of the Neepsend Steel and Tool Corporation Ltd.

▶ Mr. J. A. PEACOCK has been appointed a Director of John Lysaght's Scunthorpe Works Ltd.

▶ Mr. L. POWELL has transferred from the Research Department of the Alkali Division of I.C.I. Ltd. to the Research Department of the Company's Nobel Division, Ayrshire. He is in charge of the newly formed Materials of Construction Section.

▶ Dr. J. N. PRATT is now with the Department of Metallurgy, University of Manchester.

▶ Professor G. V. RAYNOR has been elected a Vice-President of the Institute of Metals.

▶ Mr. J. ARTHUR REAVELL, Chairman of Kestner Evaporator and Engineering Co., Ltd., London, and of Kestner (South Africa) (Pty.) Ltd., Johannesburg, is at present on a visit to the Union and Southern Rhodesia.

▶ Mr. C. W. ROBINSON, vice-consul for Sweden, has been elected President of the Birmingham Consular Association.

▶ Mr. K. SACHS has left The Mond Nickel Co., Ltd., to join the staff of the G.K.N. Group Research Laboratory, Wolverhampton.

▶ Mr. R. B. SIMS has resigned his appointment as Head of the Rolling Laboratories of the British Iron and Steel Research Association, to take up the post of Research Manager with Davy and United Engineering Company. He will remain a Consultant to the Laboratory.

▶ Mr. C. B. SNODGRASS, General Manager of Fusarc Ltd., has been appointed Director and General Manager of the Company.

▶ Mr. J. E. SRAWLEY is now on the staff of the British Columbia Research Council at the University of British Columbia, Vancouver, Canada.
▶ Mr. D. R. SWINDELL is now in the Development Section of Colvilles Ltd., Motherwell.
▶ Mr. P. H. TORTISE has left James Booth and Co., Ltd., to take up an appointment as Assistant Works Chemist at Denbro Ltd., Birmingham.
▶ Mr. R. TOYE has left the Steelmaking Division of the British Iron and Steel Research Association at Sheffield, and is now with the London Brick Co., Ltd.
▶ Mr. V. WATKINS, at present Maintenance Manager, has been appointed Deputy Chief Engineer of the Steel, Peech and Tozer Branch of United Steel Companies, Ltd.
▶ Mr. M. A. WILLIAMSON has left the Pullman Standard Car Manufacturing Co., Hammond, Indiana, to take up the appointment of Director of Research, Burroughs Adding Machine Co., Philadelphia, U.S.A.

Obituary

Mr. J. T. BANKS, of Glover Brothers (Mossley) Ltd., in June, 1952.

Mr. WILLIAM COWAN, of Wishaw, Lanarkshire, on 8th December, 1952.

Mr. FRANK L. ESTEP, of the Barium Steel Corporation, New York, on 11th December, 1952.

Dr. CHARLES H. HERTY, Jr., of the Bethlehem Steel Co., Pennsylvania, U.S.A., on 17th January, 1953.

Mr. CHARLES LESSNER, of Pontevedra, Spain.

IRON AND STEEL ENGINEERS GROUP

The **Twenty-Second Meeting** of the Group will be held at 4 Grosvenor Gardens, London, S.W.1, on **Wednesday, 18th March, 1953.** A Buffet Luncheon will be held in the Library.

The following is the detailed programme:

10.30 A.M.–12.45 P.M.—Discussion on:
"*Industrial Electric Furnaces*," by J. C. Howard (Electric Furnace Co., Ltd.) (*see* this issue, pp. 285–291)

12.45–2.0 P.M.—Buffet Luncheon

2.0–4.0 P.M.—Discussion on:
"*Electric Distribution System at the Abbey Works of the Steel Company of Wales Ltd.*," by T. B. Rolls (McLellan and Partners) and E. C. Slater (Steel Company of Wales Ltd.) (*see* Feb., 1953, pp. 163–176).

EDUCATION

National Certificates in Metallurgy

REPORT BY THE JOINT COMMITTEE ON THE PROGRESS OF THE SCHEME FOR THE AWARD OF NATIONAL CERTIFICATES IN METALLURGY FOR THE YEAR 1951–52

The Joint Committee has approved five new Schemes for National Certificates in Metallurgy during 1952; other Schemes have been revised, and Schemes are also being organized in districts not yet covered, with a view to suitable courses being started as soon as possible.

A short Refresher Course in Metallurgy, organized by the Ministry of Education for teachers of Metallurgy in Technical Colleges, was held during the Easter Vacation, 1952. The course, which dealt with modern metallurgical techniques, was very well attended and successful.

Courses approved by the Joint Committee are in operation at the following Technical Colleges:

Ordinary National Certificate—Birmingham College of Technology; Bradford Technical College; Bristol, The College of Technology; Chesterfield Technical College; Corby Technical College; Coventry Technical College; Crumlin, The Technical College of Monmouthshire; Cumberland Technical College, Workington; Derby Technical College; Enfield Technical College; Kingston Technical College, Surrey; Leeds College of Technology; Llanelly Technical College, Carmarthenshire; Manchester College of Technology; Middlesbrough, Constantine Technical College; Neath Mining and Technical Institute; Newcastle-upon-Tyne, Rutherford College of Technology; Newport Technical College, Monmouthshire; Nottingham and District Technical College; Preston, Harris Institute; Rotherham College of Technology; Scunthorpe Technical College; Sheffield College of Technology; Shotton, Deeside Technical Institute; Smethwick, The Chance Technical College; Stoke-on-Trent, North Staffordshire Technical College; Swansea Technical College; Wandsworth Technical College, London; Wednesbury, County Technical College; Wolverhampton and Staffordshire Technical College.

Contributory Centres — Bordesley Green Technical School, Moseley Grammar School Evening Institute, and Kidderminster School of Science (Contributory Centres to Birmingham College of Technology); Dudley and Staffordshire Technical College (Contributory Centre to Wednesbury, County Technical College); Nottingham People's College Senior Technical Institute (Contributory Centre to Nottingham and District Technical College); Wolverhampton, Wulfrun College of Further Education (Contributory Centre to Wolverhampton and Staffordshire Technical College).

Higher National Certificate—Battersea Polytechnic, London; Bradford Technical College; Birmingham College of Technology; Chesterfield Technical College; Coventry Technical College; Cumberland Technical College, Workington; Manchester College of Technology; Middlesbrough, Constantine Technical College; Newport Technical College, Monmouthshire; Nottingham and District Technical College; Rotherham College of Technology; Scunthorpe Technical College; Shotton, Deeside Technical Institute; Sir John Cass College, London; Smethwick, The Chance Technical College; Swansea Technical College; Wednesbury County Technical College; Wolverhampton and Staffordshire Technical College.

Final Examinations of the course for the Ordinary National Certificate have been held during 1952 at 28 Technical Colleges, and for the Higher National Certificate at 16 Technical Colleges, for candidates who have satisfied the conditions laid down in Ministry of Education Rules 111, under which the Scheme is operated.

178 candidates have qualified for the award of an Ordinary National Certificate in Metallurgy, and 103 candidates have qualified for the award of a Higher National Certificate, as a result of final examinations held in 1952. The records of a further 5 candidates for the Ordinary Certificate and 3 for the Higher National Certificate have been approved by the Joint Committee, subject to the fulfilment of certain conditions. Eight candidates for the Ordinary Certificate and 6 for the Higher National Certificate, who entered the final examinations in 1951 and whose records were approved by the Joint Committee, subject to certain provisions, have qualified for the award of Certificates in 1952. The Committee has also approved the award of the Ordinary Certificate to 8 candidates and the Higher Certificate to 17 candidates, who entered the final examinations in 1951 but whose records were not available at the date of publication of the Report for 1950–51. Distinctions have been gained by 38 candidates who have shown an

exceptional grasp of their subjects, indicating a high degree of training and knowledge in the particular subject in which the Distinction has been gained.

Comparative figures of entries and successful candidates for the seven years since the institution of the Scheme for National Certificates in Metallurgy are as follows:

Ordinary National Certificate

Year	Technical Colleges	Number of Candidates Entered	Number of Candidates Awarded Certificates
1946	4	28	23
1947	11	102	69
1948	15	123	74
1949	21	222	107
1950	20	281	142
1951	25	314	138
1952	28	334	186*

Higher National Certificate

Year	Technical Colleges	Number of Candidates Entered	Number of Candidates Awarded Certificates
1946	...	...	...
1947	2	22	18
1948	8	38	22
1949	10	74	59
1950	11	62	42
1951	16	110	54
1952	16	132	120*

Prizes were awarded to 53 successful candidates who showed particular merit in the Final Examinations in 1952, from the Prize Fund established for this purpose by The Iron and Steel Institute, the Institution of Mining and Metallurgy, and the Institute of Metals.

The Joint Committee takes this opportunity of thanking the Assessors for their valuable assistance in the operation of the scheme.

The following candidates successfully completed the senior part-time course for the year 1951–52 and have been awarded the Ordinary National Certificate in Metallurgy (a subject given in parentheses indicates the award of a distinction in that subject):

Ordinary National Certificate

Birmingham College of Technology—Barnett, Thomas Arthur; Beddoes,† Lawrence Picken; Buckley,† John Thomas; Burke,† Brian Arnold; Cox, Gordon John; Fitzsimmons,† Philip Edward; Hogg, Colin Douglas; Jackson,† Harry; Jevons,† John; Julian, Raymond Trevor; Mason,† Alfred George; Mountford,† Peter (engineering drawing); Palmer,† Colin John; Sealey,† Alan Geoffrey; Taylor, Leonard Samuel; Tranter,† Donald Frederick; Tomlinson,† Roger Charles.

Bradford Technical College—Eastwood, David Hale; McManus, Terence Patrick.

Bristol College of Technology—Damon, John David.

Chesterfield Technical College—Bassett, Alan Victor; Davies, Eric; Else, George Edwin (inorganic chemistry, physical chemistry, general metallurgy); Ferguson, John; Skevington, William Victor; Taylor, Derrick; Whitehead, Eric.

Corby Technical College—Knight, Percy William; Loasby, Frederick Anthony; Muir, John Marshall (physics); Price, Walter John C. (physics); Smith, John Edward M.

Coventry Technical College—Astfalck, Anthony Noel; Carden, Robert Leslie; Cave, Brian John; Olney, Albert Philip J.; O'Neill, John Sidney; Seal, Gerald Thomas; Young, Arthur Ernest.

Crumlin, The Technical College of Monmouthshire—Beese, John Gaythorne; Brimble, David; Hacker, Eric; Hill, Brian Samuel; Moses, Ronald George; Player, Glyn (chemistry, physics, metallurgy); Sprake, Roy Henry; Williams, William Benjamin; Williams, William Evan.

Cumberland Technical College, Workington—Burrow, Ronald; Miller, Archibald (inorganic and physical chemistry, physics, mathematics S.3); Rogers, Thomas Raymond; Sharratt, Patricia Nora; Wiggins, John.

E. Ivison and A. Miller have qualified in Mathematics S.3 for supplementary endorsement of their Certificates awarded in 1951 and 1952, respectively.

Enfield Technical College—Kellert, Bernard George; Embleton, Frederick Thomas.

Leeds College of Technology—Hill, Kenneth; Mortimer, Eric.

Llanelly Technical College—Evans, William David; Jones, John Howard; McVicar, Eric John; O'Shea, Thomas John.

Manchester College of Technology—Evans, Thomas; Graham, Joseph; Orrell, John.

Middlesbrough, Constantine Technical College—Davies, Ronald; Hardwick, Allan Terence; Malcolm, Alan; Neesam, Thomas; Simpson, Anthony Victor.

Neath Technical College—Adams, James Herbert; Bailey, Brian; Morgan, Rees; Morris, Clive; Thomas, David Alan.

Newcastle-upon-Tyne, Rutherford College of Technology—Green, Geoffrey Alan W.; Scrimgeour, Norman; Wilson, Thomas Brian.

Newport Technical College, Monmouthshire—Canham, John Daniel; Davy, Lyndon George T.; Duncan, Allan Keith; Dumayne, John Henry; Harris, John Edward; Hitchings, Ronald Frederick; Paginton, Frederick George.

Nottingham and District Technical College—Davis, John William; Johnson, Keith; Shipley, Frank Geoffrey (inorganic chemistry, metallurgy, physics).

Rotherham Technical College—Barringer, Norman Leslie; Beaumont, Jean; Blakemore, Kenneth; Bradshaw, Fred; Coulson, Geoffrey; Dowson, Derek Ralph; Fox, John Harold E.; Gill, Geoffrey Machin; Gregory, Kenneth; Guest, Cyril; Harris, Geoffrey Alan; Harrison, John; Ivey, Robert William; Jackson, Gordon Trevor; Kay, Geoffrey Hamilton; King, Ernest; Montgomery, Walter (physics); Pashley, Ronald; Priestley, John Ian; Stocks, Willis Roy; White, Carol Henry; White, Desmond; Wilson, Alan; Woodiwiss, Alec Gerald.

Scunthorpe Technical College—Atkinson, Derek William (physical and inorganic chemistry, physics); Barton, Eric (physics); Bowness, John Brian (physics); Clayton, John Brian; Jacklin, Clarence William (physical and inorganic chemistry); Jackson, Geoffrey William S. (physical and inorganic chemistry, metallurgy, physics); Johnson, Brian; Kirkby, Redvers; Long, Terence; Plastow, Byron; Reed, Alan; Shaw, Gerald Michael; Slater, Alan Thomas; Young, Dennis Russell.

Sheffield College of Technology—Brook, Raymond; Gillott, Dennis; Greaves, Peter Marsh; Gunstone, Cyril Roy.

Shotton, Deeside Technical Institute—Jones, Glyn Ivor (metallurgy); Jones, Joseph.

Smethwick, The Chance Technical College—Badham, Leslie Winnall; Bates, Dennis Ralph; Beechey, Derek Maurice; Beard, Ralph Humphrey; Beard, Samuel; Brunner, Paul Anthony; Cooper, Alan Frederick; Cooper, Frank Robert; Cornes, Gordon Henry; Denning, Dennis

* The results of 8 candidates for the Ordinary Certificate and 17 candidates for the Higher Certificate whose records were not available in 1951 have been included in 1952.

† Candidates qualified in Engineering Drawing for endorsement of their Certificates.

Frederick G.; Hickman, Donald; Nightingale, Derek Steventon.

Stoke-on-Trent, North Staffordshire Technical College—Garratt, Jean; Harrison, David Anthony; Salt, John Joseph.

Swansea Technical College—Chappell, John Brian; Francis, Walford; Harries, Howell Bevan; Hammond, Frank Sutherland; Jewell, Richard Ronald (physics); Jones, Reginald Gordon; Leyshon, David Raymond; Morgan, Don John; Rogers, David Walden (general metallurgy); Shaughnessy, Reginald Norman (general metallurgy, physics); Shorrock, James Eccles; Thomas, Eurwyn Lloyd (physics); Thomas, Idris John (physics); Williams, Ronald.

Wandsworth Technical College — Dalton, Patrick Leonard; Eldridge, George Henry; Freeman, Leonard Frank (chemistry); Harding, Alan George; Jackson, David Richard; Kelsey, Ronald Henry F.; Marshall, Robert Edward W.; Owen, Kenneth Milne; Pitcher, David Edward (general metallurgy).

Wednesbury, County Technical College (*Final Examinations*, 1951)—Baker, Philip; Bayley, Kenneth Joseph; Berry, Bert William; Challis, William Edward; Dawes, Cyril; Flavell, Maurice Harold; Jennings, Bryan.

(*Final Examinations*, 1952)—Challoner, James; Cox, Kenneth Horace; Croton, Geoffrey; Millward, Maurice James; Reynolds, Arthur Ernest; Ware, Barry (physics).

Wolverhampton and Staffordshire Technical College (*Final Examinations*, 1951)—Goodall, Raymond.

(*Final Examinations*, 1952)—Eales, Frank Alan; Martin, George Edward; Rowley, Frank.

HIGHER NATIONAL CERTIFICATE

The following candidates successfully completed the advanced part-time course for the year 1951–52 and have been awarded the Higher National Certificate in Metallurgy (a subject given in parentheses indicates the award of a distinction in that subject):

Battersea Polytechnic — Barnard, Harold William; Boughton, John Derek; Day, Robert Henry; Jarman, Raymond Arthur (metallurgical analysis); Joslin, George Herbert B.; Markson, Desmond; Martin, Edmund John; Rogers, Donald Geoffrey; Taylor, Ian Thomas G.

Birmingham College of Technology—Badham, Eric; Bancroft, Brian; Bate, Alan Thomas; Clift, Alfred Whittier (physical metallurgy); Cooper, Frank James; Evans, Andre John; Gallimore, Peter William; Hall, Dennis; Harper, Derek Genders; Murcott, Arthur William; Nation, Geoffrey; Phelps, Robert Edward; Rees, Benjamin Lewis (non-ferrous metals and alloys); Rowley, Arthur Edward S.; Rushton, Geoffrey Clive; Shaw, John Malcolm; Stanton, William Lawton; Wright, Geoffrey Henry (non-ferrous metals and alloys, metallurgical analysis).

A. W. Clift, D. G. Harper, and B. Bancroft have qualified in Metallurgical Analysis taken as an additional subject for endorsement. D. E. Hughes has qualified in Iron and Iron Alloys for endorsement of his Certificate awarded in 1951.

Bradford Technical College—Holdsworth, Fred.

Chesterfield Technical College—Bargh, Jim; Morrison, James; Stringer, Charles Roy (metallurgical analysis, physical metallurgy); Whittingham, John.

Coventry Technical College — Bromfield, Gerald; Bryant, Anthony James (mechanical working of metals); Fields, Maurice; Haynes, David Philip A.; Vivian, Gordon Philip.

Cumberland Technical College, Workington—Ashbridge, Alan; Brenan, Colin; Fearon, John Ernest; McKay, William; Nicholson, Thomas.

Middlesbrough, Constantine Technical College—Hindmarch, Philip; Hugill, Ronald William; Mansford, Ralph Ernest; Pattinson, Matthew.

Newport Technical College, Monmouthshire—Amphlett, David Charles J. (treatment of ferrous metals and alloys); Clough, William Allan; Fortey, Ronald George; Knight, Donald John; Lewis, Frederick William.

Nottingham and District Technical College—Hampson, Terence Gilbert; Millward, Cyril; Tomlinson, Gordon Raymond.

Rotherham Technical College—Almond, Francis Martin; Andrews, Alan; Beech, Frederick G.; Broadbent, Peter R.; Bunnett, Edward H.; Dickinson, Ernest Allan; Elliston, John Cooper; Hall, Raymond William; Haynes, Charles Barrie; Leeson, John; Marvill, Geoffrey; Orme, Geoffrey; Perry, Roy; Sanderson, Kenneth George; Thompson, Raymond; Topham, Rex James; Worley, Edward.

The following candidates have qualified in an additional and optional subject for endorsement purposes: F. M. Almond, J. C. Elliston, R. Perry, and R. Thompson (Metallurgical Analysis); R. W. Hall, C. B. Haynes, J. Leeson, and G. Marvill (Mechanical Testing and Heat-Treatment).

Scunthorpe Technical College—Grocock, Donald.

Shotton, Deeside Technical Institute—Ash, Arthur; Barnes, John Roy.

Smethwick, The Chance Technical College—Cooper, Alfred Ernest; Eastwood, William Brian; Fielding, Leonard John (principles of physical metallurgy, treatment of ferrous metals and alloys); Hayward, Harold Thomas (manufacture and treatment of non-ferrous metals and alloys, treatment of ferrous metals and alloys); Hickman, John St. Clair; Jones, Walter Gwynfa L.; Kent, Donald Melchoir; Kings, Geoffrey Alan; Sadler, Thomas Edwin; Simpson, William Ernest (treatment of ferrous metals and alloys); Smith, John Neville; Taylor, George Bennett (principles of metallurgical analysis, principles of physical metallurgy, manufacture and treatment of non-ferrous metals and alloys, treatment of ferrous metals and alloys); Thomas, Raymond William (principles of physical metallurgy, treatment of ferrous metals and alloys, manufacture and treatment of non-ferrous metals and alloys); Wootton, James Stanley.

A. E. Cooper, J. S. Hickman, H. T. Hayward, G. A. Kings, S. N. Smith, J. S. Wootton, L. J. Fielding, W. E. Simpson, and G. B. Taylor have qualified for endorsement of their Certificates in Treatment of Ferrous Metals. W. B. Eastwood, T. E. Sadler, and R. W. Thomas have qualified for endorsement of their Certificates in Manufacture and Treatment of Non-Ferrous Metals and Alloys; R. J. Webb has qualified in Manufacture and Treatment of Non-Ferrous Metals and Alloys for endorsement of his Certificate gained in 1951.

Swansea Technical College—Jones, Thomas Gwyn; Walkey, Ernest John W.

Wednesbury, County Technical College (*Part-time course*, 1951)—Gill, Anthony; Harper, Cecil Trevor (metallurgical analysis); Hinnett, Oliver John; Johns, Raymond; Jones, Stanley Raymond D.; Mason, John; Vaughan, Geoffrey Raymond.

J. M. Clitheroe has qualified in Treatment of Non-Ferrous Metals and Alloys for endorsement of his Certificate awarded in 1950.

(*Sandwich Course*, 1951)—Duley, Derek W.; Emmerson, Leslie G.; Gettings, George W.; Goodchild, Keith T.; Morris, Kenneth A.; Sillitoe, John F. (analysis, process metallurgy—ferrous).

The above candidates have also qualified for endorsement of their Certificates in the following additional subjects: Process Metallurgy (Non-Ferrous), Fuels,

Refractories, Applied X-rays, Mechanical Testing, Mechanical and Thermal Treatment.

(*Part-time Course*, 1952)—Sands, Raymond Leonard; Wolverson, Thomas.

R. L. Sands has also qualified in Treatment of Ferrous Metals and Alloys for endorsement of his Certificate. A. Beddow and W. F. Lycett have qualified in Extraction and Refining of Common Non-Ferrous Metals for endorsement of their Certificates awarded in 1950. S. R. D. Jones has qualified in Extraction and Refining of Common Non-Ferrous Metals for endorsement of his Certificate awarded in 1951.

(*Sandwich Course*, 1952)—Aston, Kenneth; Gwinnett, Francis Edgar; Johnson, Roy Harold (physical metallurgy, applied X-rays); Martin, Alan Howard (mechanical and thermal treatment, mechanical testing); Morris, Kenneth W.; Powell, Richard; Whittaker, Gordon.

The above candidates have also qualified for endorsement of their Certificates in the following additional subjects: Process Metallurgy (Non-Ferrous), Mechanical and Thermal Treatment, Fuels, Refractories, Mechanical Testing, Applied X-rays.

Wolverhampton and Staffordshire Technical College (*Final examinations*, 1951)—Beaman, Victor James W.; Davies, Reginald Frank; Taylor, Leslie Frank; Wood, Alan John.

(*Final examinations*, 1952)—Hadley, John F. C.; Jones, Kenneth Leslie; Lack, Bryan James; Law, Rosemary.

Prizes

Prizes have been awarded to the following candidates in respect of the Final Examinations for Ordinary National Certificates in Metallurgy for the year 1951–52:

Birmingham College of Technology—Mason, Alfred George (one guinea); Mountford, Peter (one guinea).

Chesterfield Technical College—Else, George Edwin (five guineas).

Corby Technical College—Muir, John Marshall (one guinea); Price, Walter John C. (one guinea); Smith, John Edward M. (one guinea).

Coventry Technical College—Astfalck, Anthony Noel (one guinea).

Crumlin, The Technical College of Monmouthshire—Player, Glyn (five guineas).

Cumberland Technical College, Workington—Miller, Archibald (four guineas).

Manchester College of Technology—Graham, Joseph (one guinea).

Newport Technical College, Monmouthshire—Harris, John Edward (one guinea).

Nottingham and District Technical College—Shipley, Frank Geoffrey (five guineas).

Rotherham Technical College—Ivey, Robert William (one guinea); Montgomery, Walter (one guinea).

Scunthorpe Technical College—Atkinson, Derek William (two guineas); Barton, Eric (one guinea); Bowness, John Brian (one guinea); Jacklin, Clarence William (one guinea); Jackson, Geoffrey William S. (five guineas).

Shotton, Deeside Technical Institute—Jones, Glyn Ivor (two guineas).

Stoke-on-Trent, North Staffordshire Technical College—Garratt, Jean (one guinea).

Swansea Technical College—Jewell, Richard Ronald (one guinea); Rogers, David Walden (one guinea); Shaughnessy, Reginald Norman (two guineas); Thomas, Eurwyn Lloyd (one guinea); Thomas, Idris John (two guineas).

Wandsworth Technical College—Freeman, Leonard Frank (two guineas); Owen, Kenneth Milne (one guinea); Pitcher, David Edward (one guinea).

Wednesbury County Technical College—Ware, Harry (two guineas).

Wolverhampton and Staffordshire Technical College—Eales, Frank Alan (one guinea).

Prizes have been awarded to the following candidates in respect of the Final Examinations for Higher National Certificates in Metallurgy for the year 1951–52:

Battersea Polytechnic—Jarman, Raymond Arthur (one guinea).

Birmingham College of Technology—Clift, Alfred Whittier (one guinea); Rees, Benjamin Lewis (one guinea); Stanton, William Lawton (one guinea); Wright, Geoffrey Henry (two guineas).

Chesterfield Technical College—Stringer, Charles Roy (four guineas).

Coventry Technical College—Bryant, Anthony James (two guineas).

Cumberland Technical College, Workington—Ashbridge, Alan (one guinea).

Newport Technical College, Monmouthshire—Amphlett, David Charles J. (one guinea); Lewis, Frederick William (one guinea).

Smethwick, The Chance Technical College—Fielding, Leonard John (two guineas); Hayward, Harold Thomas (two guineas); Simpson, William Ernest (one guinea); Taylor, George Bennett (six guineas); Thomas, Raymond William (three guineas).

Wednesbury County Technical College (1951)—Harper, Cecil Trevor (one guinea); Sillitoe, John F. (four guineas).

(1952)—Wolverson, Thomas (one guinea); Johnson, Roy Harold (three guineas); Martin, Alan Howard (two guineas).

Wolverhampton and Staffordshire Technical College (1951)—Wood, Alan John (one guinea).

(1952)—Lack, Bryan James (one guinea).

		£	s.	d.
Ordinary Certificate:	31 students	58	16	0
Higher Certificate:	22 students	44	2	0
	53	£102	18	0

NEWS OF SCIENCE AND INDUSTRY

Marine Steam Turbines

The Institution of Mechanical Engineers is to hold a Conference on Marine Steam Turbines on Friday 6th March, 1953, at the offices of the Institution, Storey's Gate, St. James's Park, London, S.W.1.

The Conference will be held in three sessions, Morning, Afternoon, and Evening, commencing at 10.30 A.M. The papers to be presented will review generally the work of the Parson and Marine Engineering Turbine Research and Development Association.

Flexural Plasticity in Mild Steel

The British Section of the Société des Ingénieurs Civils de France has arranged to hold a meeting on this subject at 4 Grosvenor Gardens, London, S.W.1, on Wednesday 15th April, 1953, commencing at 6.0 P.M. A paper will be presented by Monsieur A. Lazard, and the President of the Section, Mr. J. E. Swindlehurst, O.B.E., will be in the Chair. Tea will be served in the Library at 5.30 P.M.

Members of The Iron and Steel Institute are cordially invited to attend.

B.W.R.A. Summer School of Welding

The third Summer School of Welding organized by the British Welding Research Association will be held at Ashorne Hill, near Leamington Spa, Warwickshire,

from Monday 27th April to Saturday 2nd May, 1953. This year the general theme of the School is welding fabrication and production, and the various aspects will be covered in logical sequence during the course.

Those wishing to attend the School, or to send representatives from their organizations, should communicate as soon as possible with the Secretary, B.W.R.A. Summer School of Welding, British Welding Research Association, 29 Park Crescent, London, W.1.

Symposium on Spectroscopic Determination

The American Association of Spectrographers has arranged a symposium on "Emission Spectroscopic Determination of Metals in Non-Metallic Samples" in Chicago, to be held on 1st May, 1953. Contributed papers in the fields of petroleum, geology, agriculture, pharmacy, biology, ceramics, etc., are welcomed. Enquiries should be addressed to the Chairman of the Symposium Committee, Mr. J. P. Pagliassotti, c/o Standard Oil Company (Indiana), Box 431, Whiting, Indiana, U.S.A.

Third International Congress on Electrothermics

The Third International Congress on Electrothermics will be held at the Maison de la Chimie in Paris from 18th to 23rd May, 1953, under the auspices of the Comité Française d'Electrothermie.

The increasing use of electric heating in industry in the last few years has made it necessary to co-ordinate the efforts of all those concerned with research, manufacture, and application in this important new technique. The main object of the Congress will be to study the methods and applications of electric heating that have been evolved since the Second Congress, held at The Hague in 1947. Special attention will be given to economic aspects, such as the rational utilization of energy, improvement of quality, and increase in productivity.

The technical side of the Congress will be dealt with in seven sections:

1. Fusion, refining, and sintering of metals and alloys, and ore-melting
2. Heat-treatment of metals, welding, and brazing
3. High-temperature heating of non-metallic materials
4. Low-temperature heating of non-metallic materials
5. Electrochemical problems
6. Terminology, standardization, documentation, education, etc.
7. General aspects of electric heating.

Enquiries concerning the Congress should be addressed to the IIIe Congrès International d'Electrothermie, Comité Française d'Electrothermie, 2 rue Henri-Rochefort, Paris 17e.

Australian Institute of Metals

The Annual Meeting and Congress of the Australian Institute of Metals is to be held in Brisbane, Queensland, from 25th to 29th May, 1953. Technical sessions will be devoted to the presentation and discussion of papers on Chain Making and on the Production and Utilization of Rare Metals. Visits will be arranged to various industries in Queensland.

Members of The Iron and Steel Institute who may be visiting Queensland at that time are cordially invited to attend the Congress.

Errors in the A.S.T.M. Index

The A.S.T.M. Index of X-ray powder diffraction data has now been in use for about ten years, and it is recognized that it contains a number of wrong identifications which are likely to lead to confusion. The X-ray Analysis Group of the Institute of Physics has therefore proposed to collect information about the errors that have been observed by users of the Index, and to publish as soon as possible a list of all the errors that seem to be substantiated. Many, if not all, of these errors have by now probably been noted by various workers, and may in some cases have been mentioned casually in their publications, but to be of real use to others the errors should be published collectively. Everyone who has noted such errors is therefore asked, whether he has already published them or not, to send the details to Dr. A. J. C. Wilson, University College, Cathays Park, Cardiff.

Non-Destructive Testing

Solus-Schall Ltd. have opened a new non-destructive testing station at Matlock, Derbyshire. It is equipped with X-ray, gamma-ray, magnetic, and single-probe ultrasonic apparatus.

New Plant at Ilford Ltd.

Ilford Ltd. have recently installed an additional film-coating unit of modern design and a new plant, jointly owned by themselves and BX Plastics Ltd., for manufacturing film base.

United States Steel Company

On 1st January, 1953, the United States Steel Company was merged into the United States Steel Corporation, which now takes over full rights of the Company.

Steetley Magnesite Co., Ltd.

The business and assets of the Palliser Works, Hartlepool, which has been operated since 1941 as an Agency Factory and managed by the British Periclase Co., Ltd., have been acquired from the Ministry of Materials by a new company known as the Steetley Magnesite Co., Ltd., a subsidiary of the Steetley Co., Ltd., of Worksop, Notts.

Change of Address

The INDIAN INSTITUTE OF METALS has moved to new offices at 31 Chowringhee Road, Calcutta 16.

DIARY

2nd Mar.—CLEVELAND INSTITUTION OF ENGINEERS—"*Utilization of Water in Iron and Steel Works*" by W. H. Tubbs—Cleveland Scientific and Technical Institution, Corporation Road, Middlesbrough, 6.30 P.M.

3rd Mar.—INSTITUTE OF METALS—"*Recent Advances in Furnace Design*," by E. S. W. Eardley—Metallurgy Department, University College, Singleton Park, Swansea, 6.30 P.M.

3rd Mar.—NEWPORT AND DISTRICT METALLURGICAL SOCIETY—"*Flat Rolled Products*," by S. E. Graeff—Whitehead Institute, Cardiff Road, Newport, 7.0 P.M.

3rd Mar.—SHEFFIELD METALLURGICAL ASSOCIATION—"*Technological Education in Relation to the Iron and Steel Industry*," by D. R. O. Thomas—Grand Hotel, Sheffield, 7.0 P.M.

4th Mar.—MANCHESTER METALLURGICAL SOCIETY—"*Precipitation Hardening*," by H. K. Hardy—Engineers' Club, Albert Square, Manchester, 6.30 P.M.

5th Mar.—INSTITUTE OF WELDING—"*Hard Surfacing of Steel by Electric Welding*," by W. D. Biggs—

Cleveland Scientific and Technical Institution, Corporation Road, Middlesbrough, 7.30 P.M.

5th Mar.—LEEDS METALLURGICAL SOCIETY—"*Lubrication in Cold-Working Processes*," by E. McAulay—Chemistry Department, The University, Leeds 2, 7.0 P.M.

5th Mar.—WEST OF SCOTLAND IRON AND STEEL INSTITUTE (Joint Meeting with Engineers Group of The Iron and Steel Institute)—"*Experience of a Lubrication Engineer in a Steelworks*," by P. W. Murray—39 Elmbank Crescent, Glasgow, 6.45 P.M.

9th Mar.—INSTITUTION OF PRODUCTION ENGINEERS (Joint Meeting with Engineers Group of The Iron and Steel Institute and with Sheffield Society of Engineers and Metallurgists)—"*Preventive Maintenance*," by O. Lewis—Royal Victoria Station Hotel, Sheffield, 6.30 P.M.

10th Mar.—INSTITUTION OF PRODUCTION ENGINEERS—"*Case Hardening and Induction Hardening*," by W. J. G. Cosgrave—Queen's Hotel, Dundee, 7.30 P.M.

10th Mar.—SHEFFIELD METALLURGICAL ASSOCIATION—"*A New Simple Analytical Chemistry of Tantalum and Niobium based on Cellulose Chromatography*," A. F. Williams—Grand Hotel, Sheffield, 7.0 P.M.

11th Mar.—SWANSEA AND DISTRICT METALLURGICAL SOCIETY—"*The Steelworks Test House*," by A. J. K. Honeyman—Central Library, Swansea, 6.30 P.M.

11th Mar.—NORTH WALES METALLURGICAL SOCIETY—"*The Working of Metals*," by W. C. F. Hessenberg—County Primary School, Plymouth Street, Shotton, Nr. Chester, 7.15 P.M.

12th Mar.—LIVERPOOL METALLURGICAL SOCIETY—"*The Use of Aluminium in Shipbuilding and Structural Engineering*," by E. G. West—Liverpool Engineering Society, The Temple, Dale Street, Liverpool, 7.0 P.M.

16th Mar.—SHEFFIELD SOCIETY OF ENGINEERS AND METALLURGISTS (Joint Meeting with Engineers' Group of The Iron and Steel Institute)—"*New Die Block Plant at Messrs. Walter Somers Ltd., Halesowen, Birmingham*," by F. J. Somers—Mappin Hall, The University Building, St. George's Square, Sheffield, 7.30 P.M.

16th–17th Mar.—UNIVERSITY OF BIRMINGHAM (Department of Industrial Metallurgy)—Annual Conference—"*The Deep Drawing of Metals*" and "*Continuous Casting of Ingots*"—The University, Edgbaston, Birmingham 15.

17th Mar.—SHEFFIELD METALLURGICAL ASSOCIATION—"*Intergranular Corrosion of 18/8 Cr–Ni Type Stainless Steels*," by R. Butcher—Grand Hotel, Sheffield, 7.0 P.M.

19th Mar.—IRON AND STEEL ENGINEERS GROUP—Twenty-Second Meeting—Discussion on Electric Furnaces—4 Grosvenor Gardens, London, S.W.1, 10.30 A.M.

19th Mar.—SOCIETY OF INSTRUMENT TECHNOLOGY—"*The Askania Controller in the Iron and Steel Industry*," by W. C. Fahie—Cleveland Scientific and Technical Institution, Corporation Road, Middlesbrough, 7.30 P.M.

19th Mar.—STAFFORDSHIRE IRON AND STEEL INSTITUTE—"*Blast Furnace Development*," by A. Hacking—Star and Garter Royal Hotel, Wolverhampton, 7.30 P.M.

20th Mar.—NORTH EAST METALLURGICAL SOCIETY—"*The Rôle of Metallurgy in Atomic Energy*," by H. M. Finniston—William Newton School, Norton, Co. Durham, 7.15 P.M.

23rd Mar.—INSTITUTE OF METALS—43rd Annual May Lecture—"*The Present and Future Metallurgical Requirements of the Chemical Engineer*," by Sir Christopher Hinton—The Royal Institution, Albemarle Street, London, W.1, 6.0 P.M.

24th Mar.—LINCOLNSHIRE IRON AND STEEL INSTITUTE—Annual General Meeting—"*The Manufacture of Mild Steel Sheets and Strip*," by H. Edwards—Wortley Hotel, Scunthorpe, 7.30 P.M.

24th–26th Mar.—INSTITUTE OF METALS—Annual General Meeting—Park Lane Hotel, London, W.1, 10.30 A.M.

25th Mar.—EBBW VALE METALLURGICAL SOCIETY (Joint Meeting with The Iron and Steel Institute)—"*Deep Drawing*," by H. W. Swift—Lecture Room, Messrs. Richard Thomas and Baldwins Ltd., Ebbw Vale, 7.15 P.M.

27th Mar.—STAFFORDSHIRE IRON AND STEEL INSTITUTE (Associates' Section)—Annual General Meeting—Star and Garter Royal Hotel, Wolverhampton, 8.0 P.M.

27th Mar.—ROYAL AERONAUTICAL SOCIETY—Discussion on Fatigue—Ramsey Theatre, Chemistry Department, University College, Gower Street, London, W.C.1, 10.0 A.M.

27th Mar.—SOCIETY OF CHEMICAL INDUSTRY—Symposium on "*Corrosion Inhibitors*"—Chemistry Lecture Theatre, Manchester University, 10.0. A.M.

31st Mar.—SHEFFIELD METALLURGICAL ASSOCIATION—"*Stainless Steel Sheet Production*," by P. Munro—Grand Hotel, Sheffield, 7.0 P.M.

TRANSLATION SERVICE

(The previous announcement was made in the February, 1953, issue of the *Journal*, p. 188.)

TRANSLATIONS AVAILABLE

No. 462 (German). A. SEND and F. STRÄHUBER: "Determination of the Heat Losses from Blast-Furnaces by Measurements on the Cooling Water." (*Stahl und Eisen*, 1952, vol. 72, Nov. 20, pp. 1509–1512).

No. 463 (German). K. BORN and W. KOCH: "Influence of Aluminium on the Properties of Mild Unalloyed Steels." (*Stahl und Eisen*, 1952, vol. 72, Oct. 9, pp. 1268–1277).

No. 464 (German). W. HEISCHKEIL: "Influence of the Manufacturing Conditions in the Steel Plant on the Quality of Forging-Grade Steel Ingots." (*Stahl und Eisen*, 1952, vol. 72, June 5, pp. 663–668).

TRANSLATIONS IN COURSE OF PREPARATION

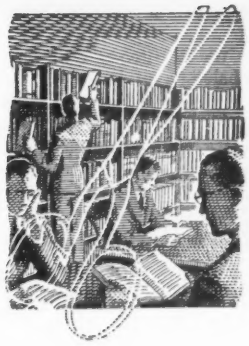
(German). F. RAPATZ and F. MOTALIK: "The Shape of Life/Cutting-Speed Curves and their Evaluation." (*Stahl und Eisen*, 1952, vol. 72, Dec. 4, pp. 1583–1587).

(German). H. BEISSWÄNGER: "The Cupping Test as a Means for Determining the Deep-Drawing Properties of Sheet and Strip." (*Metall*, 1952, vol. 6, Dec., pp. 744–753).

CHARGES FOR COPIES OF TRANSLATIONS—The charge for translations is £1 for the first copy and 10*s*. for each additional copy of the same translation. Requests should be accompanied by a remittance. These translations are not available on loan from the Joint Library.

TRANSLATIONS PREPARED AT MEMBERS' REQUESTS—Members requiring translations of foreign papers are invited to communicate with the Secretary, who will ascertain whether they can be prepared for inclusion in the Series.

ABSTRACTS AND BOOK NOTICES



IRON AND STEEL MANUFACTURE AND RELATED SUBJECTS

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ORES—MINING AND TREATMENT

Itabira Is Breaking Its Bottlenecks. J. K. Van Denburg, jun. (*Eng. Min. J.*, 1952, **153,** July, 84–89). Details are given of the mining methods and reserves of the mines of Cia. Vale do Rio Doce producing Itabira ore. Mechanized mining and ore handling has begun, and the 1951 production of $1{\cdot}3 \times 10^6$ metric tons showed a sharp rise.—K. E. J.

The Metal Mining Industry in Sweden and Norway. W. E. Selnes. (*Canad. Min. J.*, 1952, **73,** July, 57–62). Swedish and Norwegian mining practices and techniques are compared with Canadian methods. Rock-drilling, ore-transport, hoisting, and miners' welfare are highly developed in Scandinavia, but the general level of safety is not as high as in Canada. Also considered are : Shafts, crushing, hydraulic air-storage chambers, instruction and training of works and supervisors, and power supplies.—T. E. D.

Minnesota Mining Men Open up Rich Puerto Rico Iron Mine. A. W. Knoerr. (*Eng. Min. J.*, 1952, **153,** Aug., 74–79). The Keystone mine, about 32 miles from San Juan, Puerto Rico, was opened by West Indies Mining Co. in 1951, and now produces 30,000 tons/month of iron ore for shipment to the U.S.A. Stripping and mining operations are in progress : the ore shipped averages 60% Fe.—K. E. J.

Labrador Ore Loader. (*Canad. Metals*, 1952, **15,** May, 28–29). A description is given of the ore loading system being constructed at Seven Islands, Quebec. The initial annual handling capacity is to be 10 million tons rising to 20 million tons in later years. The equipment is rated at 6000 long tons/hr. when dumping ore into one of the 440,000 ton capacity stockpiles, or 8000 long tons when loading directly into ships.

Work on Taconite Now Heads Toward Commercial Goal. A. H. Hubbell. (*Eng. Min. J.*, 1952, **153,** July, 72–75). Technical and economic details are given of the pilot- and full-scale plants existing or projected for the beneficiation and pelletising of taconite iron ore in Minnesota and Michigan. Two further pilot plants will commence operations in 1952. An annual production of 10^7 tons is forecast for 1960.—K. E. J.

Tests with Magnetic Roasting of Hematite at Striberg. G. von Hofsten. (*Värm. Bergsmann. Ann.*, 1951, 58–113). [In Swedish]. An outline of roasting methods employed in the U.S.A. and Germany is followed by a description of Swedish attempts to counter a shortage of concentrates by utilising reserves of phosphoric ores (0·008 to 0·030% P). A detailed account is given of the design and performance of the vertical-shaft furnace, reduction capacity 100,000 tons/year, of operating results since 1946 and of combustion problems encountered. The furnace, 9 m. high × 4·5 × 1·5 m., has 12 inlets for producer gas. Laboratory methods of ore microscopy are discussed and a detailed account is given of the thermal and chemical conversions in the shaft, sample gas analyses and equilibrium conditions observed. Data on the heat balance, and of 1951 production costs, are presented.—G. G. K.

Simple Tramp Iron Detector Effective at Lower Cost. C. M. Marquardt. (*Eng. Min. J.*, 1952, **153,** Sept., 96–98). Simple magnetic and available electric detectors of tramp iron in ore feeds to crushers offer little protection against tough pieces of high-manganese steel. The 'Detramp' is a robust apparatus, easily maintained, designed for this purpose. The tramp steel produces a small current in passage through a magnetic field, and this is stepped up to 60-cycle A.C., amplified, and passed to a warning system. Sensitivity can be varied according to size or type of tramp material.—K. E. J.

Sulphur Elimination. R. A. Elliot. (*Canad. Min. J.*, 1952, **73,** July, 51–56). The history of sintering is surveyed. Tests with a 75–90 lb. batch sintering pot are described, and a pilot-scale Dwight-Lloyd sinter strand is mentioned. Tests were made with high-sulphur hematite and high-sulphur siderite, using the contained sulphur as fuel. The limit of the amount of sulphur fuel that can be used before the residual sinter sulphur increases, is 6·0–9·0% with hematite, and 6·0–7·5% with siderite. An amount of 2·65 g. of FeS_2 burned to Fe_3O_4 is equivalent to 1 g. carbon. The use of sulphur as fuel is considered possible, but slower sintering rate means less production per unit.—T. E. D.

The Development of the Siegerland Preparation and Calcination Techniques, and Problems Still Awaiting Solution. H. Gleichmann. (*Metall u. Erz*, 1944, **41,** July, 145–149). An historical survey is made of the development of modern preparation and calcination methods for the Siegerland manganiferous iron ore deposits, with special reference to wet-mill and electromagnetic concentration, and the flotation of sulphides of copper, lead and zinc. Improvements over the last 20 years are outlined. Small and large-scale experiments have been carried out on the enrichment of low-grade sludges, with partial success.—L. H.

FUEL—PREPARATION, PROPERTIES, AND USES

Fuel Control at Port Talbot, Margam and Abbey. J. T. Davies. (*Iron Coal Trades Rev., Special Issue*, 1952, **165,** Oct. 1, 116–120). The author discusses the fuel requirements of the Port Talbot, Margam, and Abbey plants of the Steel Company of Wales Ltd., and outlines the functions of the fuel control room.—G. F.

Investigation of the Firing of Lower-Silesian Carbonaceous Shale in a Mechanized Shaft Furnace. S. Pawlowski. (*Prace Inst. Metalurgii*, 1952, **4**, 3, 243–260). [In Polish].

Fuel Requirements of the Future: The Place of Coal. D. Hicks. (*Iron Coal Trades Rev.*, 1952, **165**, Sept. 12, 569–575 ; Sept. 19, 637–646). In discussing future fuel requirements, the author deals first with the sources and outputs of primary forms of energy (*e.g.* coal, oil) and the conversion of these to secondary forms (*e.g.*, gas, electricity). The consumption of energy by the main consumers is next considered. Our future energy supplies will have to be drawn mainly from our rapidly diminishing and unrenewable coal reserves, and the author calls for an accurate estimate of the quantity and quality of these reserves, contending that Britain's survival as an industrial nation depends on our ability to provide a sufficiency of energy at a competitive price.—G. F.

Proposals to Improve the Basis of Efficiency Calculations on the Preparation of Mineral Coal. K. Grumbrecht. (*Glückauf*, 1952, **88**, Sept., 957–964). Methods of calculating yields of washed coal, and coal and dirt mixtures are discussed. A new method is developed and its advantages explained by comparison with current methods. A nomogram, incorporating the method of least squares, is shown and its use explained. The specific gravity method of sample analysis is considered, and methods for determining the probable errors are dealt with.—P. F.

Drainage Behaviour and Water Retention Properties of Fine Coal. D. W. Gillmore and C. C. Wright. (*Min. Eng.*, 1952, **4**, Sept., 886–894). In view of the present interest in recovery of coal fines from colliery effluents and refuse banks, a reproducible laboratory technique (analogous to plant conditions) has been evolved to study gravity drainage variables. A column of wet coal essentially reaches an equilibrium condition (with saturation, transition, and low-moisture zones) after 24 hr. Results are quoted for the effects of particle size, size distribution, bunker design, and wetting agents on drainage behaviour.—K. E. J.

Comparative Effectiveness of Coal Cleaning Equipment. O. R. Lyons. (*Min. Eng.*, 1952, **4**, Sept., 895–902). A correlation method, based on the amount of displaced material and the difficulty of separation, has been developed to assess the effectiveness and capacity of coal cleaning plants. The method is applicable to any type of equipment, and 105 sets of plant data are analysed. The results confirm quantitatively many recognized comparisons. (30 references).—K. E. J.

Coke Plants Explosion-Proofed. C. F. Stissel. (*Steel*, 1952, **131**, July 7, 103–120). The principles involved in fire- and explosion-proofing coal, coke, and by-product plants are discussed. Segregation of hazards, explosion-proof electrical fixtures, cleanliness, proper maintenance, and drainage are factors of safety.—A. M. F.

Coal Preparation, Coke Ovens, and By-Product Plant. H. J. Victory. (*Iron Coal Trades Rev., Special Issue*, 1952, **165**, Oct. 1, 60–66). Details are given of the new and existing items of the coal preparation, coke-oven, and by-product plant of the Margam and Abbey Works of the Steel Company of Wales Ltd.—G. F.

The Banded Constituents of Coal and Their Agglutinizing Curves. E. P. Reim. (*J. Chem. Met. Min. Soc. S. Africa*, 1952, **52**, Jan., 156–160). Brief mention is made of the four banded constituents of coal, and agglutinizing curves are given for each.—D. H.

Constitution of Coal and the Formation of Coke. W. A. Kirkby and R. J. Sarjant. (*Nature*, 1952, **170**, Oct. 11, 597–599). Four coal slacks of similar chemical composition, but different coking properties, have been examined by solvent extraction with chromatic absorption, and by infra-red absorption spectrometry. The latter enables the coals to be grouped in relation to coking properties.—A. G.

Investigations on the Application of a Coal Charge of High Bulk Density for the Production of Metallurgical Coke. F. Byrtus and A. Foerster. (*Prace Inst. Metalurgii*, 1952, **4**, 3, 213–222). [In Polish]. The influence of compressing various coal charges on the mechanical properties of the coke produced and the output of a coke-oven was investigated. The tamping time was taken as a measure of the degree of compression. Experiments were also performed in a small oven with a charge composed of briquetted coal. Satisfactory results were obtained. An optimum charge density was found which made possible at least a 3% increase in the output with a simultaneous improvement in coke quality.—V. G.

New Coke-oven Plant Doubling Output at Clyde Iron Works. (*Iron Coal Trades Rev.*, 1952, **165**, Sept. 19, 619). A description is given of the new coke-oven and by-product plant installed at the Clyde Iron Works of Colvilles Ltd. The new ovens will produce 6000 tons of blast-furnace coke and 600 tons of domestic coke per week, thereby doubling the existing output. The by-product plant contains a number of innovations, which are enumerated.—G. F.

The By-Product Coking Industry in Canada. E. J. Burrough. (*Canad. Min. J.*, 1952, **73**, Sept., 63–73). The Canadian by-product coking industry is surveyed in its historical, technical, and economic aspects. Recent coke-oven installations are mentioned, and the standard by-product coke-oven, the Curran-Knowles coke-oven, and the travelling grate carbonization process are very briefly described. Selection of coal and coke quality are considered, and reserves of coking coals are given. The by-products are listed and dealt with in some detail.—T. E. D.

St. Louis Coke Ovens. (*Coke Gas*, 1952, **14**, Sept., 303–310). The coke-oven battery built by Simon-Carves Ltd. for the Great Lakes Carbon Corp. at St. Louis, Missouri, is described. It is the first battery of British design and construction to be erected in the U.S.A. The battery consists of 40 ovens of the compound underjet type, with twin compartment regenerators, which are designed for heating with either coke-oven or producer gas. Details are given of the battery and also of the by-product plant, for which additional units were ordered ; these are now being put into operation.—T. E. D.

Acid Resisting Construction in By-product Plant. R. Nicklin. (*Coke Gas*, 1952, **14**, Sept., 323–324). Resinous cements are now available for lining most types of chemical plant in conjunction with acid-resisting bricks or tiles ; these can in many cases be used with advantage as substitutes for lead. Acalor No. 9a, acid- and alkali-resisting impermeable cement is used successfully in sulphate of ammonia plants and in a wide range of chemical plant and pickling tanks.

Investigates Methods for Capturing Dusts in Gases. S. Vajda and G. M. Dreher. (*Steel*, 1952, **131**, Aug. 4, 105–120). Pilot tests conducted at a steelmaking shop, for controlling the amount of dust in gases exhausted from the stocks, are described. Gases were drawn out of the stack with a movable scoop, through a gas-to-water heat exchanger, by an exhaust fan, and discharged into the cleaning unit being tested. Units tested included a sonic agglomerator, a Rotoclone, a cyclone-type tube collector, a high-voltage electrical precipitator and a pebble-type filter. It is concluded that a high-voltage electrostatic precipitator is the only satisfactory method of reducing the amount of dust.—A. M. F.

Cleaning Industrial Gases. J. J. Seaver. (*Iron Steel Eng.*, 1952, **29**, Oct., 84–91). The author describes the dry, wet, and electrical precipitator systems for cleaning industrial gases. Dust catchers, baffles, deflectors, dry electrical precipitator and ' bag house ' for dry systems, and sprays, gas scrubbers, fans, disintegrators, eliminators, and wet electrical precipitators are examined for the wet cleaning method. Technical data and a number of layout arrangements are given.

Industrial Gas Development. (*Coke Gas*, 1952, **14**, Sept., 328–331). Some advances in the use of gas in manufacturing processes are described and discussed. They include the Hypact burner (which has been adapted for preheating rolls in the tinplate industry), paint curing by radiant heat, and the production of a low-sulphur gas for use in the preparation of metallurgical furnace atmospheres, and other industrial processes.—T. E. D.

TEMPERATURE MEASUREMENT AND CONTROL

Optical Temperature Measurements and a Contribution to the Radiation Analysis of Cast Iron. K. Orths. (*Stahl u. Eisen*, 1952, **72**, Oct. 23, 1349–1353). It has been shown by comparative measurements on foundry irons with the Bioptix colour pyrometer and a total radiation pyrometer that, as is the case with steel melts, the total radiation pyrometer is often completely unsuitable for accurate measurement of the temperature of molten cast iron. The simultaneous determination of the black body and true temperature enable some predictions regarding quality to be made so that, by fixing the most favourable casting temperature and carrying out a radiation analysis, a considerable reduction in numbers of rejects can be brought about.—J. P.

Measuring Steel Temperatures in the Basic Converter with an Immersion Thermocouple. K. G. Speith and O. Därmann. (*Stahl u. Eisen*, 1952, **72,** Oct. 23, 1336–1346). A mobile platinum/platinum-rhodium immersion pyrometer, the method of calibration, and accuracy are discussed. The effect of temperature at various periods during refining on castability, and on the nitrogen, silicon, and phosphorus contents, has been determined. Methods, including oxygen enrichment of the blast, have been studied for controlling temperature so as to produce better quality steel.—J. P.

Temperature Standardization: Big Gains in Research. R. E. Wilson. (*Steel*, 1952, **131,** Aug. 11, 90–92). Research efforts at the American National Bureau of Standards are being directed toward the extension of the international temperature scale to provide greater accuracy and reproducibility in the measurements thus made possible. Pressure control has been extended to measurements of temperature at the sulphur point, a noise thermometer is being developed to measure thermodynamic temperatures, and tests are being carried out with an iridium/iridium-rhodium thermocouple for extremely high temperature measurement.—A. M. F.

REFRACTORY MATERIALS

Chromite Investigations Part III.—Variations in the Composition of the Pure Chromite Mineral from the Eastern Chrome Belt, Lyderburg District. J. F. De Wet. (*J. Chem. Met. Min. Soc. S. Africa*, 1952, **52,** Jan., 143–155). A number of chromite samples, representative of all the types existing in the Eastern Chrome Belt, were taken according to a systematic scheme of sampling. The pure chromite mineral was extracted from these by suitable ore-dressing techniques and analyzed for their chromic, ferric, and ferrous oxide contents. An assay of the chromic oxide content of the sampled ore was taken in some cases. The results are tabulated to illustrate the variations that occur and the different chromites are classified according to their composition.—D. H.

Modernizing the Production of Silica Sand. (*Eng. Min. J.*, 1952, **153,** Sept., 104–106). Several unique features have been incorporated into the processes of preparation of dune sands for use in the glass, ceramic, building, and foundry industries, by the Del Monte Properties Co. A flowsheet is shown. After conversion of the sand into pulp (75% solids), it is scoured and scrubbed in Wemco six-cell attrition machines without damage to the individual grains. Further treatment is by flotation processes instead of gravity methods.—K. E. J.

How to Lengthen Refractory Life. A. H. Thomson. (*Canad. Metals*, 1952, **15,** Feb., 24–25). The author discusses refractories for basic electric steel furnaces and how to get the best results with present-day materials. Roof life may be improved by care in heating and cooling, no over-charging, and avoidance of flux additions in the powdered form. The composition of the slag should be carefully controlled to avoid attack on the walls. Fettling refractories and basic mortars with added bonds should be carefully selected.—J. C. B.

New Fibre Filters, Insulates—Stands up at 2300° F. (*Iron Age*, 1952, **170,** Aug. 21, 133–134). A new synthetic refractory fibre which has high heat resistance, light weight, and low heat transmission has been introduced by the Carborundum Co. It is being used for high temperature insulation in combustion and exhaust systems of jet engines. Further applications suggested are for gas and fume filtration, industrial furnace insulation, and to replace asbestos.—A. M. F.

BLAST-FURNACE PRACTICE AND PRODUCTION OF PIG IRON

Thoughts on the Production of Iron. R. Durrer. (*Von Roll Werkzeit.*, 1952, **23,** Mar., 33–36). In 1925 the iron production of the world passed the 100 million ton mark. In 1950, the 200 million mark was reached. In 1950 and 1951 about 250 million tons of ore and more than 100 million tons of coke were consumed. If the present trend in production continues, the resulting intensification of the raw materials problem will soon lead to the displacement of industries, and to basic changes in production methods.—E. C. S.

Wet Dust Recovery: Big Tonnage Returns. M. L. Cover. (*Steel*, 1952, **131,** Aug. 11, 98–109). The method of wet dust recovery from blast-furnaces and its conversion into filter cake for sintering is discussed. The cake should have 27 to 30% moisture and should constitute about 25% of the total sinter mix. The pumps used for the washer water should be rubber lined, and single tray filter units are preferred.—A. M. F.

Blast Furnace Practice. C. E. Agnew. (*Steel*, 1952, **130,** Mar. 17, 98–110 ; Mar. 24, 90–94 ; Mar. 31, 86–95 ; Apr. 7, 121–138 ; Apr. 21, 98–104 ; Apr. 28, 84–90 ; May 19, 106–127 ; June 2, 102–121 ; June 9, 106–122 ; June 16, 142–146 ; June 23, 82–89). Continuation of a series of articles, *see J. Iron Steel Inst.*, 1952, **172,** Oct., 234. This series of articles covers the generation, recovery, and use of heat, the comparative values of fluxing stores in the south and east areas, a review of slag composition research, the raw materials problem and its effect on blast-furnace development, the benefication of raw materials, and blast treatment including the use of oxygen.

The Desulphurization of Pig Iron with Acid Slags. H. G. Erne. (*Roll'schen Eisenwerke Mitt.*, 1952, **11,** Jan.–June, 1–70). Present-day theories of desulphurization are reviewed. Experiments show that at 1700° C., the gaseous desulphurization of pig iron can exceed 60%. All factors which tend to reduce the oxygen partial pressure tend also to promote desulphurization. Thus a good gaseous desulphurization requires (1) low oxygen content of the pig-iron phase, (2) a very acid slag rich in silicon, and (3) a high temperature. These conditions will, however, result in an undesirable concentration of silicon in the molten pig. This can only be avoided by treatment in a vacuum at a relatively low temperature.

Margam Blast Furnaces. R. Frost. (*Iron Coal Trades Rev., Special Issue*, 1952, **165,** Oct. 1, 67–74). The author gives details of the ore handling and preparation plant, the blast furnaces, and auxiliary plant of the Margam Works of the Steel Company of Wales Ltd.—G. F.

DIRECT PROCESSES

On the Catalytic Decomposition of Carbon Monoxide by Iron Reduced Electrolytically at Low Temperatures. B. Fleureau and A. Sancelme. (*Compt. Rend.*, 1952, **235,** Oct. 13, 801–803). A comparison of the catalytic effect of ferric oxide reduced at various temperatures between 1200° and 275° C., and of ferric oxide electrolytically reduced, shows that the latter causes rapid carbon deposition between 400° and 600° C.

PRODUCTION OF STEEL

Notes on the Visit to the Metallurgical Industries of the United States. A. Peters. (*Rev. Tech. Luxembourg*, 1952, **44,** July–Sept., 155–170). An account is given of a visit to a number of American steel producing centres. Several recommendations are made for improving European practice.

First Operating Experiences at Margam and Abbey Works. W. F. Cartwright. (*Iron Coal Trades Rev., Special Issue*, 1952, **165,** Oct. 1, 58–59). The author outlines the first operating experiences of the new Abbey Works and reconstructed Margam Works of the Steel Company of Wales Ltd., and quotes recent production figures.—G. F.

Canada's Steel Industry: Plans to Increase Capacity. (*Iron Coal Trades Rev.*, 1952, **165,** Oct. 3, 733–735). This article briefly outlines the development of the Canadian iron and steel industry and describes its present plant. The completion of expansion schemes will increase the production of pig irons by 45%, of steel ingots by 30%, and of rolled and drawn products by 30%, above the 1950 levels.—G. F.

Luxembourg Iron and Steel Industry: Post-War Developments and Position under Schuman Plan. (*Iron Coal Trades Rev.*, 1952, **165,** Sept. 19, 659–660). This article outlines post-war developments in the Luxembourg iron and steel industry, and indicates its present capacity. The effects of the Schuman plan on Luxembourg's industry are considered.

Spectral Control of the Basic Bessemer Converter. R. Breckpot, B. Juchniewicz, and C. de Clippeleir. (*Rev. Mét.* 1952, **49,** Aug., 552–560). From the simultaneous measurement of intensities in two monochromatic ranges (6300 Å and 4300 Å), curves of relative intensity against converting time are plotted. These reveal the transitions between decarburization, dephosphorization, and iron oxidation. If blowing is stopped when indicated by these curves, phosphorus is below 0·04%, provided that the bath temperature is below 1580° C. Higher bath temperatures lead to greater loss of iron by oxidation and lower the phosphorus content.—A. G.

Research on the Deoxidation of Open-Hearth Killed Steels. M. Nepper and H. Herbiet. (*Rev. Univ. Mines*, 1952, Series 9, **8,** July, 274-280). It is shown that it is possible to evaluate the quality of the killed steel on the basis of the total oxygen content and the content of certain oxidised inclusions. It is possible to manufacture, in the open-hearth furnace, steel

containing <0·005% oxygen and having excellent macroscopic properties. The size of the austenite grains increases with the residual aluminium contents. Some notes on the methods of oxygen analysis used are given.—B. G. B.

Activities of Constituents of Iron and Steelmaking Slags. Part I—Iron Oxide. E. T. Turkdogan and J. Pearson. (*J. Iron Steel Inst.*, 1953, **173,** Mar., 217–223). [This issue].

Steel Plant Operations at Port Talbot, Abbey, and Margam Works. R. W. Evans. (*Iron Coal Trades Rev., Special Issue,* 1952, **165,** Oct. 1, 75–81). The author describes the layout and plant of the Abbey, Margam, and Port Talbot melting shops of the Steel Company of Wales Ltd., and outlines the steelmaking operations.—G. F.

Converts Melt Shop to All-Electric. (*Steel,* 1952, **131,** Aug. 25, 110). A brief description is given of three new 80-ton top-charged basic electric furnaces installed in the Timken Roller Bearing Co.'s works at Canton. The furnaces are to be charged by two sizes of clam-shell-bottom buckets (950 and 450 cu. ft.).—A. M. F.

Control of Demand and Power Factor on Electric Arc Furnaces. H. J. Heath. (*Iron Steel Eng.*, 1952, **29,** Oct., 78–81). The author considers the variables which the electric arc furnace operator has to contend with and shows that, if he is supplied with the proper instruments and trained in their application, furnace operation will be considerably simplified and improved.—M. D. J. B.

Primary Crystallization of Hadfield Steel. J. Haas. (*Prace Inst. Metalurgii,* 1952, **4,** 3, 177–212). [In Polish]. Published data and current theories on the primary crystallization of metals are reviewed. The author's view with regard to steel is that the lowering of the interphase temperature and the increase in the maximum temperature of nucleation are the factors which stop further growth of the columnar zone. The first of these factors predominates in small ingots and thin-walled, rapidly cooled castings ; the second factor predominates in large ingots and thick-walled, very slowly cooled castings. In intermediate cases both factors act simultaneously. (101 references).—V. G.

PRODUCTION OF FERRO-ALLOYS

Manganese and Titanium Spark Revival in Southern Nevada. J. B. Huttl. (*Eng. Min. J.*, 1952, **153,** Sept., 124–126). At Henderson, Nevada, the properties and plant of a large war-time magnesium unit are being adapted for work on manganese and titanium. Combined Metals Reduction Co. will produce ferromanganese in electric furnaces from manganese ore nodules, Manganese Inc. will treat ore from a nearby mine by dressing and calcining to produce manganese ore nodules. Titanium Metals Corp. will produce ductile titanium ingots by the Kroll process.—K. E. J.

FOUNDRY PRACTICE

How One Foundry Meets Management Problems. G. Barrett, R. J. Young, and J. H. Walsh. (*Canad. Metals,* 1952, **15,** Apr. 21, 26–29). This is a brief description of the steel foundry of Dominion Engineering Works, Ltd., Montreal. The normal castings required are given with the equipment available. Moulding and control practices are described as well as a brief section on inter-departmental co-operation. The overall productivity of the foundry is given as about 80 man-hours per net ton.—J. C. B.

New Foundry Techniques. (*Canad. Metals,* 1952, **15,** Jan., 28-30 ; Feb., 26–27). Research and developments discussed at the Technical and Operating Conference of the Steel Founders' Society of America are reviewed. Hydroblast sand reclamation is shown to be efficient and economical. Flame gouging and cleaning are represented by the oxy-acetylene, Arcair and Kwik-arc Jet Torch processes. High-purity oxygen is claimed to increase cutting rates in the foundry, and also cleaning costs are reduced by the pneumatic edge grinder. Among the new equipment discussed are a combination skid and shake-out basket, and two shockless jolt machines. Floating cores are quoted as being practical and reducing eventual cleaning costs.—J. C. B.

Oxygen Enriched Cupola Blasts, R. O. Day. (*Canad. Metals,* 1951, **14,** Dec., 22–27). This is a report on tests conducted by Linde Air Products Co. It is claimed that oxygen enrichment decreases the time required to bring the cupola up to temperature, decreases the coke rate, and permits lower quality coke to be used. It does not affect the lining nor change the pattern of cupola operations, and the installation cost of equipment is low. Oxygen consumption varies, but an average figure was 13 cu. ft./1000 cu. ft. of wind for each 1% enrichment.—J. C. B.

Notes on the Manufacture of Castings Containing Spheroidal Graphite. T. Boissaux. (*Rev. Tech. Luxembourg,* 1952, **44,** Apr.–June, 82–88). A review of methods of making nodular iron is given. The theories proposed to explain the production of spheroids are discussed. The elements capable of inducing the process are viewed with particular emphasis on cerium and magnesium.—B. G. B.

A New Low-Cost Process for the Industrial Production of Nodular Iron. C. Longaretti and M. Noris. (*Fonderia Ital.*, 1952, **1,** June, 3–9). This paper examines a new method of preparing nodular iron, based on the use of porous inclusions of pure magnesium. These inclusions are kept immersed in the bath until they are completely absorbed. The result of numerous tests confirm that the method can be applied to all types of iron and that production costs are considerably reduced.—M. D. J. B.

The Manufacture of Large Gear Wheels in Spheroidal Graphite Iron. C. Galletto. (*Fonderia Ital.*, 1952, **1,** Aug., 58–62). This article describes the manufacture of large gear wheels in cast iron containing spheroidal graphite obtained by addition of magnesium. Details are given of the foundry equipment, which is capable of casting gears of up to 5 tons.

Malleable Cast Iron Resistant to Abrasion. I. Khoroshchev. (*Przeglad Odlewnictwa,* 1952, **2,** 7–8, 258–263). [In Polish]. Malleable cast iron of pearlitic or pearlitic–ferritic structure can be used as a substitute for bearing alloys. The results of the investigations made by the Institute of Agricultural Machinery of U.S.S.R. concerning the making and utilization of this iron are described.—V. G.

On the Question of Design Strength of Malleable Iron Castings. E. Mickel. (*Giesserei,* 1952, **39,** Sept. 4, 429–431). Malleable iron comes between steel and cast iron, as far as design strength is concerned ; whilst its fatigue strength is of the order of 15–17 kg./sq. mm.; its notch and shape sensitivity is moderate ; there is no significant difference in the fatigue strength of cleaned and polished test pieces and cleanly cast specimens. These conclusions are based on the author's experimental investigation, particulars of which are given.

The Sesci Furnace. P. C. Fassotte. (*Brit. Cast Iron Res. Assoc., J. Res. Development,* 1951, **4,** Aug., 17–20). Seven practical examples of melting practice in various sizes of Sesci horizontal rotary furnaces, ranging from 2 to 6 tons in capacity, are given. Two of the examples refer to whiteheart malleable practice and the remainder to blackheart malleable.

Experimental Production of Nodular Graphite in Cast Iron. A. I. Krynitsky and H. Stern. (*Foundry,* 1952, **80,** Mar., 106–111, 243–246 ; Apr., 98–99, 241–249). The authors describe the effect of magnesium, magnesia, magnesite, iron sulphide, calcium, nickel–chromium–molybdenum, and boron on the graphite structure of cast iron. Completely nodular graphite occurred only with iron with more than 0·03% magnesium. Two types of structure were observed—radial and amorphous. Rapid cooling promoted nodule formation, though this effect was less with more magnesium present. The study supports the theory that nodular graphite develops during solidification.—E. T. L.

Some Effects of Magnesium on the Formation of Graphite in a Solidifying Cast Iron. I. C. H. Hughes. (*Inst. British Foundrymen Preprint No.* 1036, June, 1952 ; *Foundry Trade J.*, 1952, **93,** Sept. 25, 349–356). Previous explanations of the mechanism of spheroidal graphite formation are examined. Microscopical and cooling-curve studies show that solidification occurs by the initial precipitation of primary austenite dendrites followed by the growth of graphite and austenite at the eutectic temperature. Flake graphite grows in contact with the liquid during solidification, whereas each spheroidal graphite particle develops an austenite envelope. When a magnesium-containing iron is not inoculated its eutectic temperature is 13° C. lower.—E. T. L.

Ductile Iron—Its Significance to the Foundry Industry. A. P. Gagnebin. (*Foundry,* 1952, **80,** June, 128–131, 226–230, 232–240). This illustrated article shows the range of applications for ductile iron, and traces the history of its development. The effect of the alloying elements used in its production on the strength, machinability, and resistance to wear and oxidation are discussed.—E. T. L.

The Manufacture of Metal Patterns and Core Boxes by the Clay-Plaster Method. A. Menden. (*Giesserei*, 1952, **39,** July 10, 346–347). The principles of this technique are illustrated by examples, including the valve housing of a diesel engine.

Effects of Annealing on the Removal of Burned-on Sand. D. Alverson. (*Amer. Foundryman*, 1952, **21,** May, 91–92). Annealing breaks down the surface structure of pearlite and steadite into the less brittle ferrite, so that runners of the latter are not so easily removed.—E. T. L.

Gates and Heads for Steel Castings. J. H. Hall. (*Foundry*, 1951, **79,** Nov., 108, 109, 236–248 ; Dec., 104–109, 192 ; 1952, **80,** Jan., 102–105, 237–242 ; Feb., 98–99, 212–220 ; Mar., 112–115, 257, 258 ; Apr., 104–107, 255–257). The general principles of design and application of sprues, gating systems, risers, and feeding techniques are described. Examples are given of stepped gates for high castings, swirl gates, and of partial and complete reversal of the mould to avoid difficulties inherent in top pouring. In the latter, the casting is gated at the bottom. The so-called French gate for casting plates about 2 in. thick is described. The author discusses techniques for overcoming shrinkage by pipe eliminators, the use of large feeder heads and blind risers in shell casting, and shows how to determine the size of feeder heads for sound castings. The Washburn or necked-down head, centre-line shrinkage, and the use of padding and metal chills to secure directional solidification are also dealt with.—E. T. L.

Finding the Most Profitable Snagging Wheel for the Job. P. E. Kiefer. (*Amer. Foundryman*, 1952, **22,** July, 32–34). Tests were made on a range of grinding wheels for cleaning castings, measuring the electrical energy consumed, the ratio of work to wheel wear, and the labour cost. The sum of these three factors gave a cost for each pound weight of metal grindings, for each wheel, thus allowing a cost comparison. The softest wheel is not always the best cutting wheel, but ordering hard wheels is false economy since it slows up the grinding.—E. T. L.

Moulding Machines and Their Application. W. A. Cleary. (*Australian Found. Trade J.*, 1952, **3,** Apr., 3, 5, 7–11, 13). The main types of machine are described, and the reasons that make each suitable for particular classes of work are explained.—P. M. C.

HEATING FURNACES AND SOAKING PITS

Soaking Pit Covers Poured with Castable Refractory. W. B. Smith. (*Steel*, 1952, **131,** July 21, 100–103). Details are given of a monolithic soaking-pit cover lining made with a castable refractory material with a suitable anchor assembly. The material cost is higher than firebrick, but less than preburned special shapes. Longer service life is said to offset the additional cost.—A. M. F.

Avoiding Trouble in Soaking Pit Operation. D. J. McGeary. (*Iron Steel Eng.*, 1952, **29,** Oct., 92–94). This article gives a brief review of the soaking-pit practice at the Steel Company of Canada Ltd. and shows how causes of delay have been removed. Brief reference is also made to equipment maintenance.—M. D. J. B.

Radiant-Type High Speed Furnaces for Forging Efficiency. (*Canad. Metals*, 1951, **14,** Dec. 28–29). A radiant type, gas-fired, Selas forging furnace installation at Locke Incorporated, is described. It is a continuous furnace for heating bars for forging. The high-speed heating, maintained by electronic control, gives increased forgeability, reduces scale and lengthens die life.—J. C. B.

Industrial Electric Furnaces. J. C. Howard. (*J. Iron Steel Inst.*, 1953, **173,** Mar., 285–291). [This issue].

Direct Resistance Heating. L. B. Reed. (*Steel*, 1952, **131,** July 14, 86–87). The application of this method of heating is briefly discussed. Advantages of the method are that no expensive machinery is required, and it is not limited by shape, mass, or electrical conductivity. Approximate methods for determining the power requirements are given.

HEAT-TREATMENT AND HEAT-TREATMENT FURNACES

Better Heat-Treating Born in Research. C. L. West. (*Iron Age*, 1952, **170,** Aug. 14, 134–136). A brief description is given of the work being carried out in the laboratories of the Electric Furnace Co. Salem, Ohio. The research is intended to develop operating procedures that can be readily duplicated in the field. Work on continuous heat-treating methods has included the collection of data on heat absorbtion rates, annealing and normalizing temperatures, time at temperature, cooling rates through the lower critical ranges, and the time element for cooling to the low discharge temperature required in the production of strip steel for tinplate.—A. M. F.

One Way to Increase Heat-Treating Furnace Efficiency. F. Panlener. (*Mat. Methods*, 1952, **36,** Aug., 106–107). The production of a particular furnace was considerably increased by redesigning the trays so that more components could be carried in the same space.—P. M. C.

How to Set Up a Heat-Treating Department. W. G. Patton. (*Iron Age*, 1952, **170,** Aug. 7, 109–113). A description is given of a recent installation dealing with 150–175 tons of small parts per week. Equipment includes two carburizing furnaces, three salt baths, a tempering furnace, oil and caustic quench baths, and a cooling tower. Many carbon and alloy steels are treated and case depths can be held within 0·004–0·005 in.

Selection of Steel for Automobile Parts. A. L. Boegehold. (*Metal Progress*, 1952, **62,** Sept., 67–71). The heat-treatment of alloy steels for use in automobile parts is reviewed. The use of **S**-curves, and of charts derived from them, to predict rates of cooling (such as occurs during quenching) is discussed.

New Continuous Normalizing Plant: Heat-Treatment of Steel Plates from 7-ft. Mill at Appleby-Frodingham. (*Iron Coal Trades Rev.*, 1952, **165,** Sept. 19, 650–653, 660). This article describes and illustrates the new continuous normalizing plant which has been installed at the Appleby-Frodingham Steel Co. to meet the increased output of 2800 tons/week of the 7-ft. plate mill. Some of the problems of design and location encountered during the installation are discussed.

Testing Round Carbon Drill Steel. P. L. Russell. (*Min. Eng.*, 1952, **4,** Jan., 42–43). A progress report is presented of trials by the U.S. Bureau of Mines concerning heat-treatment of the shank ends of drill steel. Drilling tests were made in greenstone and epidosite. Results indicate that soaking of drill steel during tempering does not necessarily result in a metallurgical notch or cause early failure. The effects of variations in chromium and nickel content require further examination.—K. E. J.

Mechanism of the Carburization of Some Stainless Steels. J. B. Giacobbe. (*Amer. Soc. Met. Preprint* No. 3, 1952). A carburization theory suggests that, in alloy steels, solid carbon cannot supply diffusible carbon. For nascent carbon to be generated there must be an 'oxidizing potential' caused by the formation and subsequent dissociation of carbon monoxide. In the experimental investigation, it is shown that certain stainless steels can absorb carbon from graphite in a low-pressure atmosphere. The rate of carburization by this mechanism is greatly affected by the pressure of contact between the metal and graphite.—E. T. L.

Recent Developments in Case Hardening. J. Winning. (*Mech. World*, 1952, **132,** Oct., 150–153). Box and gas carburizing are discussed in general terms. Attention is paid to compositions of carburizing mixtures, fine-grained steels, and the chemistry of carburization is discussed in some detail.

Austempered Lawnmower Blades are Hard, Tough. W. G. Patton. (*Iron Age*, 1952, **170,** July 31, 88–90). The blades (of 0·6 and 0·7% carbon steel) are treated in salt baths after forming. The hardening temperature is 1550–1600° F. and the transpformation temperature is 600° F. The hardness is RC 48 to 52, and the blades exhibit high ductility and good toughness.—A. M. F.

Flame Hardening. N. H. Cuke. (*Canad. Metals*, 1952, **15,** Mar., 62–64). A description of the process of flame hardening is given. Its advantages include the ability to harden surfaces of thin sections, control of heat input to avoid overheating or excessive grain growth on the metal surface, and, using the correct rates of cooling, great flexibility in the hardness produced. An important economic advantage is that the equipment is portable, so that large and heavy parts to be hardened need not be moved.—J. C. B.

Flame Hardening for Economic Heat-Treatment. E. F. Green. (*Canad. Metals*, 1951, **14,** Nov., 36–40). A description of the process is given, with several applications. To obtain the best results, the carbon of the steel should be over 0·35% (preferably 0·45–0·50%), although alloying elements may influence this. The method can be applied to grey iron parts. For good reproducibility, the process should be made mechanical, using either a special flame torch or the oxy-acetylene torch. Hardening temperatures are usually between 760 and 1000° C. and the work is then quenched in

air, water or oil. Finally, stress-relieving at 150–200° C. is carried out.—J. C. B.

FORGING, STAMPING, DRAWING, AND PRESSING

An Episode in the History of the British Drop-Forging Industry. L. Aitchison. (*Metal Treatment and Drop Forg.*, 1952, **19,** May, 205–210, 216). The advent of World War I and the rapid growth of aeronautics brought many problems to the drop-forging industry. The author helped to form a travelling commission representing steelmakers, drop-forgers, and inspectors, to examine all the troubles that had arisen. An account, based on the commission's experience, is given of the difficulties of that period.—P. M. C.

Application of Electric Upset Forging. P. Granby. (*Steel Processing*, 1952, **38,** May, 228–232, 243). The principles and applications of this novel method of upsetting are described and illustrated. Bar stock is slowly upset between anvil electrodes, the necessary heat being supplied by passage of electric current from one anvil to the other through the portion of bar being upset. The great advantage is that the bar stock can be automatically moulded progressively through the clamping anvil, and a large upset portion, up to 30 times the bar diameter, can thus be produced. The exact amount of material necessary to fill a subsequent drop-forging die impression can thus be obtained in one operation. Applications of the method are outlined, and available machines described.—P. M. C.

Induction Heating for Bolt Heading. (*Metal Treatment and Drop Forg.*, 1952, **19,** May, 227, 228). Birlec Ltd. have recently developed a range of standardized induction-heating machines for bolt 'pins.' Stock sizes ranging from $\frac{1}{2}$ in. to $1\frac{1}{4}$ in. in dia. can be accommodated, and the rate of delivery of heated pins can be adjusted up to a maximum output of 1500/hr./machine.—P. M. C.

Thicker Boiler Plate Formed Faster. (*Steel*, 1952, **131,** Aug. 11, 84–86). **New Press Pierces, Bends Boiler Parts.** (*Iron Age*, 1952, **170,** July 31, 96–97). **Hydraulic Press and Drawbench.** (*Mech. Eng.*, 1952, **74,** Oct., 818–820). **Babcock and Wilcox Installs New Forming Equipment.** (*Iron Steel Eng.*, 1952, **29,** Aug., 143–145). This article describes a new vertical press and a horizontal drawbench for producing hollow forgings and for bending heavy plate at the Barberton, Ohio, plant of the Babcock and Wilcox Co. Brief technical details are given of the equipment which will produce seamless hollow forgings from 8 to 26 in. inside dia. and $\frac{3}{4}$ to $4\frac{1}{2}$ in. wall thickness in lengths up to 22 ft.—M. D. J. B.

Automatic Forging Presses Feature New Shell Line. J. A. Verson and H. Irwin. (*Iron Age*, 1952, **169,** May 22, 117–125). A fully automatic production line for producing shells by the hot extrusion method is in use at the General American Transportation Co. The line comprises automatic hot-forging, extrusion, and sizing presses with pneumatic transfer equipment. Other features are hydraulic bolster pads on the presses and water-cooled die pots and punches. Shell concentricity is held within 0·015 in. at high production rates.—A. M. F.

Hot Bending Large Diameter Pipes on Site. (*Welding and Metal Fab.*, 1952, **20,** July, 246–247). A description is given of the bending of large (6–20 in. in dia). pipes on site as in the construction of oil refineries.—U. E.

Composite Forgings Cut Tooling Costs. (*Overseas Eng.*, 1952, **25,** July, 426–427). A section of tool steel (which can be fully machined beforehand) is flash-butt welded to a mild-steel base which has been previously flame-cut to the required shape. Several examples of the construction of these composite blanking dies for use in the automobile construction industry are given.—B. G. B.

Works of the German Art Iron Worker Fritz Kühn. W. Nagel. (*Ossature Métallique*, 1952, **17,** Apr., 181–185). This is an illustrated description of some artistic wrought iron work by the contemporary artist F. Kühn, based on his books "*Geschmiedetes Eisen*" and "*Sehen und Gestalten Natur, und Menschenwerk.*"—P. F.

Iron and Steel and the Forging of Works of Art. F. Kühn. (*Stahl u. Eisen*, 1952, **72,** June 19, 766–773). The growth and decline of the making of iron gates, balustrades, and other artistic work is discussed. The development of modern craftsmanship in this field is treated at length and illustrated by many photographs.—J. P.

Forge Shop Safety is Good Business. R. H. Larson. (National Safety Council, 39th Congress: *Steel Processing*, 1952, **38,** June, 286, 287, 299). Many hints and tips are given concerning plant layout, hammer maintenance, mechanical handling, and safety precautions which help to make the author's plant more efficient.—P. M. C.

Preventive Maintenance In Forge Shops. S. H. Janeski. (National Safety Council, 39th Congress: *Steel Processing*, 1952, **38,** May, 232, 233, 257). This is a general appraisal of the benefits to plant, output, and operators' welfare, which derive from planned routine inspection and maintenance.

Developments of Cold-Flow Pressing and Extrusion of Steel. K. Sieber. (*Machinery*, 1952, **80,** May 22, 891–897; May 29, 937–941). In Part I the fundamentals of cold-forming processes are considered and suitable work materials for cold-flow pressing are discussed. In Part II the field of application of the cold-flow pressing of steel, tools for cold-flow pressing, and the reinforcement of press dies are discussed.—E. C. S.

The New Pittsburgh Plant of Fisher Body. J. C. McComb. (*Steel Processing*, 1952, **38,** June, 271–276). This two-year-old plant, which produces motor-body panels for General Motors Corp., is described and illustrated. There are 79 major presses ranging from 300 to 1100 tons capacity, and 90 smaller ones ranging from 32 to 130 tons capacity. Production flow, scrap handling, and maintenance are discussed.

New Stamping Process Developed at Douglas Aircraft. T. A. Dickinson. (*Steel Processing*, 1952, **38,** June, 277–279). Preliminary details are given of the newly developed Wheelon process, which is claimed to combine the low cost advantages of the Guérin process with the production virtues of older stamping techniques. The method is based on a new type hydraulic press, capable of exerting three times the pressure normally used for aircraft parts, although it is only one-tenth the size of a conventional press of the type usually associated with the Guérin process.—P. M. C.

Self-Centralizing Tensioning Strip Guides for Piercing Dies. (*Sheet Metal Ind.*, 1952, **29,** May, 427–430). The article gives details and diagrams of simple, adjustable-tension and automatic equipment.—P. M. C.

Stainless Gutters, Downspouts, Easily Fabricated. (*Iron Age*, 1952, **169,** June 12, 126–127). The fabrication of type 430 stainless steel into roof drainage fittings by rolling, pressing, and forming can be done on the same equipment as for mild steel. Soldering with a stainless steel flux can be readily carried out, but thorough final cleaning is most important.—A. M. F.

Press-Forging Machines. (*Overseas Eng.*, 1952, **26,** Sept., 46–47). A new range of press-forging machines manufactured by Wilkins and Mitchel Ltd. is described; the capacity of the presses ranges from 600 to 3000 tons.—B. G. B.

Saving Material by Cold Forming. J. R. Fawcett. (*Mech. World*, 1952, **131,** Apr., 168–170). A general review is made of methods of saving material by the application of machines which manipulate rather than remove metal.—D. H.

Better Design Permits Cheaper Stampings. F. Strasser. (*Iron Age*, 1952, **169,** May 8, 113–115). Design for stampings should be such that there is the minimum, or if possible, no scrap left from the stock. A few simple modifications to achieve this are given. Large workpieces can often be made more economically in two or more parts.—A. M. F.

Calculating Presswork Pressures. (*Machine Shop Mag.*, 1952, **13,** Aug., 358–360). The operation of a new electronic computing instrument is described; it accurately determines the minimum tonnage rating of presses required for deep drawing and other operations.—B. G. B.

Heat Exchangers Require Precise Fabrication. F. A. Guba. (*Iron Age*, 1952, **170,** Aug. 21, 127–129). The factors affecting successful tube expanding are quality and characteristics of the tube and tube sheets, preparation of both tube and sheet, workmanship, and physical properties of the steel used. The tubes should have uniform dimensions and physical properties throughout their length; the tube sheet holes should have the correct tolerances, and the rolling in should be done with controlled torque. Steel with a low yield point is desirable.

Problems in Connection with the Drawing of Steel Wire. C. O. von Hofsten. (*Värm. Bergsmann.*, 1951, 32–53). [In Swedish]. Patenting gives a mixture of fine lamellar pearlite and bainite with good cold-working properties. If the period in the cooling bath is insufficient, residual austenite will change to martensite, giving a brittle wire. Diagrams show carburization and decarburization among test pieces annealed simultaneously under producer gas at 680° C. It is stressed that gas composition must be closely balanced. A section on

protective gases discusses disadvantages of ordinary producer gas, the high carbon-content causing temperature variations. The use and manufacture of Monogas is described. The use of HCl or H_2SO_4 for pickling is discussed and details are given of the Du-Pont sodium-hydrate bath and the Hooker NaOH + $NaNO_3$ bath.—G. G. K.

ROLLING-MILL PRACTICE

Operating Experience with Regulating Systems for the Steel Industry. W. R. Harris. (*Iron Steel Eng.*, 1952, **29,** Oct., 113–124). The author describes the four types of regulating equipments available for heavy-duty rolling-mill applications, namely : Electro-mechanical, electronic, and rotating regulators, and magnetic amplifiers. A critical examination is made of these four types and technical and operating data are given, as well as circuit diagrams.—M. D. J. B.

Rectifier Applications on Steel Mill Drives. G. B. Scheer. (*Iron Steel Eng.*, 1952, **29,** Oct., 107–109). The author describes the installation of rectifiers at the Fontana rolling mills of H. J. Kaiser Co., and shows how one of the greatest advantages of rectifiers is the overload capacity. Even when rectifiers are back, this only means short interruptions and a minimum loss of production time.—M. D. J. B.

Thoughts on Rolling-Mill Rolls Made of Chilled Cast Iron. T. Boissaux. (*Rev. Tech. Luxembourg*, 1952, **44,** July–Sept., 171–180). The various methods of heat-treatment of rolling mill rolls are discussed. The advantages of a structure of sorbite-martensite (obtained by using alloying elements) over the normal pearlite structure are explained. The particular influence which a number of elements (C, Si, Mn, S, Ni, Cr, Mo) have on the hardness of the roll is discussed. The author concludes with an analysis of the properties of a number of alloys used for American rolls.—B. G. B.

Small Two-High Reversing Mills for the Rolling of Semi-Finished Products. E. T. Peterson, L. W. King, and E. C. Peterson. (*Iron Steel Eng.*, 1952, **29,** Oct., 69–76). The authors review the trends in mill design over recent years and conclude that the simpler and more effective 2-high reversing mill has come to stay. The 3-high mill of simple design has such serious drawbacks with regard to quality and flexibility of production and cost of operation that it becomes impractical for modern installation.—M. D. J. B.

Atlas Builds First Stainless Strip Mill. (*Canad. Metals*, 1952, **15,** Jan., 14). The first strip mill in Canada for the production of stainless steel strip will be erected by Atlas Steels Ltd. at Welland, Ontario. The mill will roll continuous strip in coils in gauges from 8 to 30. A standard Sendzimir mill will reduce the coils of strip from the hot mills after their annealing and cleaning.—J. C. B.

Cold Rolling and Treatment of Steel Strip. L. Gascuel. (*SIM la Documentation Métallurgique*, 1952, Jan.–Feb.–Mar., 13–24). A review is made of different types of rolling-mill equipment used for steel strip, and of the different methods of heat-treatment used to anneal the product.—B. G. B.

Ferblatil. A Modern Cold Rolling and Electrolytic Tinning Plant. M. Gevers. (*Rev. Univ. Mines*, 1952, Series 9, **8,** June, 212–228). A description of this new plant is given, including details of the rolling mills and drives, the degreasing equipment, and the electrolytic tinning plant.—B. G. B.

Civil Engineering at Trostre Works. A. J. Williams. (*Iron Coal Trades Rev., Special Issue*, 1952, **165,** Oct. 1, 10–16, 49). The author discusses the civil engineering aspects of the contruction of the Tostre Works of the Steel Company of Wales Ltd., dealing mainly with choice of site, piling, excavation, and construction of foundations and buildings.—G. F.

Rolling and Processing at Trostre Works. J. S. Gazard and G. E. Jones. (*Iron Coal Trades Rev., Special Issue*, 1952, **165,** Oct. 1, 17–31). This description of rolling and processing at the Trostre Works of the Steel Company of Wales Ltd. includes details of continuous pickling, acid recovery and regeneration, the five-stand tandem mill, electrolytic cleaning lines, heat-treatment, preparatory and cutting-up lines for coils, electrolytic tinning lines, and hot-dip tinning plant.—G. F.

Electrical Installations: Trostre Works. J. C. Louth. (*Iron Coal Trades Rev., Special Issue*, 1952, **165,** Oct. 1, 32–35, 38). This article describes the electrical supply and distribution, the main electrical drives and controls, and instrumentation at the Trostre Works of the Steel Company of Wales Ltd.

Use of Rolls at Abbey and Trostre Works. S. E. Graeff. (*Iron Coal Trades Rev., Special Issue*, 1952, **165,** Oct. 1, 53–57). The author discusses the requirements of the rolls used in the modern four-high mills of the Abbey and Trostre Works of the Steel Company of Wales Ltd., dealing mainly with the hot-strip and cold-reduction mills.—G. F.

Operation of the Rolling Mills and Accessory Plant at Abbey Works. H. H. Ascough. (*Iron Coal Trades Rev., Special Issue*, 1952, **165,** Oct. 1, 82–94). A description is given of the operation of the slabbing and 80-in. hot-strip mills of the Abbey Works of the Steel Company of Wales Ltd., and of the subsequent treatment given to the strip, sheet, and plates produced.—G. F.

MACHINERY FOR IRON AND STEEL PLANT

Mechanical Services. G. E. Turpin. (*Iron Coal Trades Rev., Special Issue*, 1952, **165,** Oct. 1, 36–38). The author gives additional data now available on a number of the mechanical services at the Trostre Works of the Steel Company of Wales Ltd.—G. F.

Metallurgical Department: Trostre Works. W. J. S. Roberts. (*Iron Coal Trades Rev., Special Issue*, 1952, **165,** Oct. 1, 39–42). The author describes the organization of the metallurgical department of the Trostre Works of the Steel Company of Wales Ltd., and discusses its main functions of routine control, metallurgical and chemical services, and development work.—G. F.

Central Maintenance Worskhops. A. Henderson. (*Iron Coal Trades Rev., Special Issue*, 1952, **165,** Oct. 1, 95–97). A description is given of the layout of the central maintenance workshops and of the services which they provide for the Abbey, Margam, and Port Talbot Works of the Steel Company of Wales Ltd.—G. F.

Power Plant at Margam and Abbey Works. W. P. C. Ungoed. (*Iron Coal Trades Rev., Special Issue*, 1952, **165,** Oct. 1, 107–110, 115). A description is given of the various new boiler and turbine plants and of the water-treatment plant at the Margam and Abbey Works of the Steel Company of Wales Ltd.—G. F.

Electrical Installations at Port Talbot, Margam, and Abbey. J. H. Groocock. (*Iron Coal Trades Rev., Special Issue*, 1952, **165,** Oct. 1, 102–106). The author describes the electrical supply and distribution, the electrical machines and their control and maintenance at the Port Talbot, Margam, and Abbey Works of the Steel Company of Wales Ltd.—G. F.

Realigning Crane Runways in an Open-Hearth Building. M. F. Reese, jun. (*Iron Steel Eng.*, 1952, **29,** Sept., 93–99). This article describes the realignment of the crane gantries in the melting shop of Bethlehem Steel Co. at Sparrows Point. The gantries had settled from their original position in a manner which interfered with open-hearth operations and resulted in excessive maintenance on the stage and casting side cranes, due to the twisting and racking of the main shop structure. The author describes the method of re-aligning and jacking up the stanchion bases.—M. D. J. B.

Crane Runway Maintenance. H. R. Knust. (*Iron Steel Eng.*, 1952, **29,** Sept., 84–91). The author shows that, by having an adequate and well-organized building repair organization, employing a good inspection programme, making the correct analysis of each particular problem, and following through with the proper planning of each job, the difficult and often neglected work of crane runway maintenance can be successfully accomplished.—M. D. J. B.

Crane Wheel Tread Contours. J. J. Stolz. (*Iron Steel Eng.*, 1952, **29,** Oct. 101–104). The author describes a number of crane wheel tread contours and discusses the good and bad features of each. He does not recommend a particular type of wheel tread.—M. D. J. B.

Luxembourg Steelworks Plant with a Dual Function. (*Oil Eng. and Gas Turb.*, 1952, **20,** Mid-Oct., 227). A brief description of a 5400-kW. Brown Boveri gas turbine is given. This is a single-shaft recuperative plant operating on blast-furnace gas and driving a generator. An important feature of the installation is that compressed air for process duties is bled off from a point beyond the compressor delivery—a steady fall in electrical output being experienced as the quantity of air extracted is increased. This enables the gas turbine to act as a stand-by to the blast while engaged on its primary duty of generating power.—B. G. B.

Heat Transfer from Tube to Water in a Condenser. J. Jennings. (*Mech. World*, 1952, **132,** Oct., 161–165). A useful explanation is given of the physical principles involved.

The author develops a simplified formula and develops a nomogram for facilitating the design calculation.—D. H.

Metallurgical Tips for the Maintenance Man. M. V. Herasimchuk. (*Iron Steel Eng.*, 1952, **29**, Sept. 73–80). The author presents a number of basic principles underlying steel mill failures and makes recommendations for dealing with such problems. The first job of the maintenance man is to collect pertinent information about the location and type of failure. Design and metallurgical factors are distinct in their effects, yet often overlap. Relatively minor design or metallurgical changes will produce effective results.—M. D. J. B.

Training Maintenance Personnel. W. R. Fisher. (*Iron Steel Eng.*, 1952, **29**, Sept., 177–181). This article describes the maintenance training programme adopted by the Armco Steel Corp.—M. D. J. B.

Productive Electrical Maintenance. A. D. King. (*Iron Steel Eng.*, 1952, **29**, Sept., 173–176). The author suggests that maintenance should be less a service and more a part of production. Only by this means will stoppages on mills be reduced to a minimum. The need for scientific approach to the problem is stressed.—M. D. J. B.

A.C. Power Distribution in Steel Plants. D. L. Beeman. (*Iron Steel Eng.*, 1952, **29**, Sept., 140–149). The author covers primary A.C. power distribution systems and considers the nature of steel mill loads, selection of system voltage, utilization of voltages, 13·8 *versus* 6·9-kV. power distribution systems both for small and large power systems, over-all system cost differences, circuit layouts, use of system voltages above 15 kV., selection of transmission voltage above 15 kV., use of the load centre principle, grounding, and voltage regulations.—M. D. J. B.

Modern Machine Tools for the Steel Plant. W. J. Pearson. (*Iron Steel Eng.*, 1952, **29**, Sept., 182–188). Steel plant production methods are discussed ; these are not on a mass production basis as in the motor car, ratio and household appliance industries. This requires that special machine tools be as universal in use as is possible without reverting to the regular machine-shop tools. Several tools are described ; amongst these are an automatic cold saw, duplicating equipment on lathes, template attachments for use when turning mill rolls, railway axle cutting and turning equipment, and duplicators on vertical millers.—M. D. J. B.

Reconditioning Air Filter Oils. G. Findlay. (*Iron Steel Eng.*, 1952, **29**, Sept., 160–164). After briefly describing the advantages of using filtered air for cooling large motors and motor generator sets, and for use generally in motor rooms, the author discusses ore filters and the reconditioning of oils. Reference is made to an air filter unit which has operated under severe conditions for 3½ years without loss of operating time for repairs and cleaning.—M. D. J. B.

LUBRICANTS AND LUBRICATION

Selection of Seals for Various Duties. E. P. Stahl. (*Machinery*, 1952, **81**, Aug. 28, 367–373). One of the most important factors in securing optimum service from bearings, whether of the anti-friction or the plain sleeve type, is the selection of the correct means of sealing and protecting them from dirt and contamination. Among the seals considered are simple felt seals which act as oil reservoirs, metal labyrinth seals, unit spring-loaded leather seals, unit synthetic rubber seals, springless unit seals, and mechanical seals.—E. C. S.

Lubrication of Steel Surfaces Treated in a Sulphurous Bath. F. Morel. (*Compt. Rend.*, 1952, **235**, July 28, 284–286). A friction couple applied to the surfaces under a normal load of up to 100 g. is balanced by a torsion bar. With treated surfaces there is a marked lessening in friction variations and decreasing wear. Also the friction coefficient is independent of speed of surface movement.—A. G.

Research and Development at Fulmer Hall. (*Scientific Lubrication*, 1952, **4**, July, 12–14). This is a short account of the Central Development Department of the Development Division of Monsanto Chemicals Ltd., which opened at Fulmer in 1949. The description concentrates on work on additions for lubricating oils (for petrol and diesel engines, turbines, and general lubrication), particularly engine screening tests.

Methods of Testing the Quality and Economy in Use of Emulsifiable Oils, for Drilling, Cutting, Grinding, and Drawing. (*Österr. Masch. Elektrowirtschaft*, 1952, **7**, June, 213–215.)

Applications of Bentone Lubricants. D. H. Compondu. (*Iron Steel Eng.*, 1952, **29**, Sept., 165–166). Bentone lubricants are not true greases in that the oils are not thickened by a soap, but are gelled by a chemical compound obtained by reacting a purified bentonite and an organic base. The preparation of these lubricants, which are exceptionally stable over a wide range of uses, is described. The applications of these lubricants in steelworks are dealt with.—M. D. J. B.

Standardization in the Lubrication Department. J. D. Lykins. (*Iron Steel Eng.*, 1952, **29**, Sept., 155–158). The author discusses the general problem of standardizing lubrication methods and lubricants in iron and steel works. It is claimed that standardization of lubricants and equipment has (*a*) reduced unit buying costs by 10%, (*b*) simplified the training of operatives, and (*c*) reduced inventory and storage space.—M. D. J. B.

Steel Mill Lubrication—Past, Present and Future. N. I. Whiteley. (*Iron Steel Eng.*, 1952, **29**, Sept., 152–154). The author describes modern methods of lubricating and equipment for lubrication in iron and steel works, with particular reference to blast-furnace plant and electric overhead travelling cranes.—M. D. J. B.

New Developments in Lubrication Practice. C. M. Schaeffer. (*Iron Steel Eng.*, 1952, **29**, Sept., 106–108). The author shows how standard types of lubricating and allied equipment can be applied in order to gain complete automatic control of the various lubricants required throughout a steel mill.—M. D. J. B.

Wire Rope Lubrication. C. H. Layton. (*Iron Steel Eng.*, 1952, **29**, Sept., 167–171). The author stresses the need for wire rope lubrication and distinguishes between lubrication while the wire rope is being made and the proper lubrication of the wire rope in service, and describes how these operations are carried out.—M. D. J. B.

Gear Lubrication Symposium. (*J. Inst. Petroleum*, 1952, **38**, Aug., 606–698 ; Sept., 699–804). A symposium on gear lubrication, organized by the Institute of Petroleum, was held in London on Feb. 13, 1952. The papers in Part I, on the lubrication of gears, were entitled.:

Hydrodynamic Theory in Gear Lubrication. A. Cameron. (614–622).

Gear-Tooth Wear. G. I. Finch and R. T. Spurr. (623–624).

Criteria Governing Scuffing Failure. F. T. Barwell and A. A. Milne. (624–632).

Some Factors Affecting Gear Scuffing. H. D. Mansion. (633–645).

The Effect of Variation of Viscosity with Pressure on the Load-Carrying Capacity of the Oil Film between Gear-Teeth. E. McEwen. (646–650).

Note on Performance of Graphited Oil. E. A. Smith. (650–652).

The Wear and Pitting of Bronze Disks Operated Under Simulated Worm-Gear Conditions. L. S. Evans and R. Tourret. (652–668).

Part II covered the testing of gear lubricants, and it comprised the following papers :

The Development of a Test Method for Gear Lubricants on the I.A.E. Machine, by the Mechanical Tests of Lubricants Panel of Standardization Sub-Committee No. 5. (705–711).

From Test Machine to Gear-Box: Problems Associated with the Translation of Laboratory Test Results into Predictions of Field-Service Performance of Gear Lubricants. J. R. Hughes. (712–718).

Road and Laboratory Wear Tests of Gear Oils. H. L. Bingham and J. G. Withers. (718–728).

Production of Lubricant Test Gears. R. D. B. Stone. (728–732).

The Testing and Performance of Lubricants for High-Speed Gears. J. Greenwood and R. W. Morton. (732–746).

Developments in the Specification and Testing of Lubricants for Hypoid Axles with Particular Reference to a Full-Scale British Test Procedure. A. Towle. (747–762).

The Testing and Selection of Gear Lubricants. H. J. Watson. (763–774).

Service Tests in Rear Axles of Buses and Trolley-Buses. H. E. Styles and A. T. Wilford. (775–782).

WELDING AND FLAME-CUTTING

Sigma Welding Stars on 3 New Jobs. F. Harkins. (*Machinist*, 1952, **96,** Oct. 4, 1617–1619). Speeds up to 10 times those previously possible are being obtained at Solar Aircraft Co., San Diego, Calif. on difficult welding by shielded inert-gas metal-arc welding (referred to as 'Sigma welding') in three fields in which this welding with consumable electrodes has until now had only limited application.—E. C. S.

Effect of Vibration on Weld Metal. J. W. Welty. (*Welding J.*, 1952, **31,** Aug., 361s–366s). The effect of vibration on the microstructure and on mechanical properties of stainless steel types 321 and 347 steel during arc and resistance welding was investigated. There was little consistency in the purely qualitative data.—U. E.

Porosity in Mild Steel Weld Metal. D. Warren and R. D. Stout. (*Welding J.*, 1952, **31,** Aug., 381s–386s). This is a review of literature on each of the factors affecting porosity in mild-steel weld metal. (39 references).—U. E.

Evaluation of the Circular Patch Weld Test. J. E. Hockett and L. O. Seaborn. (*Welding J.*, 1952, **31,** Aug., 387s–392s). The applicability of the circular-patch weld test as a means of determining weldability was investigated. The type of specimen used has high biaxial residual stresses which are prone to cracking. Four series of unnotched Union-melt welded patches and one set of manually welded patches were bent at different temperatures to determine what effects welding variables had on the transition temperature. The transition temperatures of Union-melt welded specimens were lower than that of the manually welded series. The energy to fracture, maximum bend angle, and maximum deflection were markedly affected by welding variables.—U. E.

Production of High Quality Tubes by Electric Resistance Welding. (*Engineering*, 1952, **174,** July 18, 65–67). **Steel Tube Production by Electric Resistance Welding.** (*Engineer*, 1952, **194,** July 18, 94–95). This article describes two tube-making plants at the Corby Works of Stewarts and Lloyds Ltd., for the production of high-quality boiler and superheater tubes. The electric resistance welding process described presents a significant advance in this country. Technical and operational details are given and the measures taken to ensure high quality and reliability are described.

Resistance Welding Quality Control. T. J. Lepito and R. M. Taylor. (*Welding J.*, 1952, **31,** Aug., 692–698). Resistance welding control which embraces surface finish, chemical analysis, and thickness of the alloys for jet-engine fabrication is discussed; in addition, machine settings and testing of completed welds are important for obtaining acceptable resistance welds.—U. E.

Ceramic Backing for Tube Joints. B. Ronay. (*J. Amer. Soc. Naval Eng.*, 1952, **64,** May, 359–369). High pressure steam piping is regarded as the most critically stressed component of a steam power plant. Most joints have to be made away from workshops, and this handicaps the proper alignment of metal backing rings. Experience shows that the greatest difficulty in pipe welding, using metal backing rings, is the development of sound root conditions. The use of non-metallic backing in the joints defined in this paper offers a solution to the most objectionable feature of pipe welding.

17 Cr Stainless Replaces 18-8 in Many Weldments. G. E. Linnert. (*Iron Age*, 1952, **169,** June 26, 97–100; July 3, 128–131). Steels with 17% Cr, when properly handled, show as good corrosion resistance as 18/8 steel in welded parts. Ferrite grain growth and martensite will reduce joint properties, but can be controlled. Both this type and a titanium-stabilized type show less distortion in welding. Gas torch welding should not be used as grain growth and carbon pick-up may occur. Preheating when using electrodes minimizes the pinhole conditions sometimes encountered.—A. M. F.

Nodular Cast Iron Welds Hard to Machine. E. F. Davis and E. L. Layland. (*Steel*, 1952, **130,** June 23, 78–79). An investigation has been made into the arc welding of nodular iron and the effects of preheating and post-annealing on the soundness of the joint. The weld can approach the strength of the parent metal, but lacks ductility. This hardness is accounted for by martensite and flake graphite in the weld zone; annealing after welding, and better welding techniques, may improve the ductility.—A. M. F.

Welding Rods for the Welding of Cast Iron. (*Stanki i Instrument*, 1951, No. 4, 35). [In Russian]. The production and properties of welding rods for welding grey cast iron are briefly described.—S. K.

Ferritic Welding Electrodes Save Alloy, Give Better Welds. Z. J. Fabrykowski. (*Iron Age*, 1952, **170,** July 24, 93–97). Savings in strategic alloys of about 85% have been obtained by using ferritic welding electrodes instead of austenitic electrodes by the American Army Ordnance. The practice is being used on U.S. military vehicles on thickness up to $\frac{5}{8}$ in. where ballistic requirements must be met. The electrodes are low in hydrogen.—A. M. F.

Influence of Hygroscopic Moisture of Coatings upon some Properties of Welding Electrodes. E. D. Lonskii. (*Avtogennoe Delo*, 1951, **22,** 12, 17–19). [In Russian]. The capacity of electrode coatings to adsorb moisture and the influence of hygroscopic moisture on some properties of welding electrodes were investigated.—V. G.

Characteristics of Covered Welding Electrodes. W. Hummitzsch. (*Schweisstechn.*, 1952, **6,** May, 49–57). The physico-chemical and metallurgical properties of covered welding electrodes are discussed.—U. E.

Automatic Welding with Contact Electrodes. W. P. van den Blink, H. Bienfait, and J. A. van Bergen. (*Philips Tech. Rev.*, 1952, **13,** Mar., 247–253). An automatic welding machine, which uses contact welding rods, is described in detail. A number of welding heads, each carrying an electrode in contact with the workpiece, are mounted at appropriate intervals along a rail. Each electrode takes over the arc from its predecessor as soon as the latter is consumed. In this way, welding of any length can be done in mass production with unskilled labour. Very good uniform welds are obtained, the joints between the beads being scarcely perceptible.—P. M. C.

Investigation on the Weld Hardness of Structural Steel. E. Folkhard. (*Schweisstechn*, 1952, **6,** Aug., 85–88). The hardness of weld metal and adjacent zone of welded structural steel is determined by means of the Vidser's hardness test and H. Hauttmann's 'Rolldur' test in which a ball $\frac{1}{16}$ in. in dia. is rolled at a speed of 0·25 mm./sec. under a load of 10 kg. over the material to be tested. The Brinell hardness can be read from a scale which gives this from the width of the track.—R. A. R.

Welding in the Soviet Union. J. Mannin. (*Engineers' Digest*, 1952, **13,** Oct., 351–354). The author critically reviews the remarkable developments in welding which have taken place in the Soviet Union since 1945.—R. A. R.

Production Problems—XVII: Broken Spring Clips. (*Iron Steel*, 1952, **25.** July, 319–320). A description is given of an investigation into the causes of an excessive number of breakages encountered in the spot-welded zone of steel spring clips after a barrelling operation. Variations in the welding procedure and the use of a steel of too high carbon content are found to be the main causes.—G. F.

Welding of Heat Resistant Steels. H. F. Tremlett. (*Welding Metal Fab.*, 1952, **20,** July, 258–261; Aug., 299–303). The author discusses various aspects of heat-resistant steels as they affect the engineer. The resistance to scaling and the effect of alloying elements such as chromium, nickel, and titanium upon the properties of the steels concerned are considered. Problems arising when welding heat-resisting steels are dealt with and certain precautions are advocated.

Experiences Gained from Examinations for Vessel and Pipe Welders. K. Schmitz. (*Schweissen u. Schneiden*, 1952, **4,** Aug., 290–292).

The Performance of Hard Facing Welding Alloys. N. D. Berrick. (*Australasian Eng.*, 1952, June 7, 54–59). Elementary mechanism of wear is considered, and laboratory and field methods of measuring wear are discussed. Different types of hard-facing welding electrodes are described, and the manner in which they may be applied to reduce wear is indicated by reference to the following specific examples: Power house stoker ram boxes, internal combustion engine valves, power shovel bucket teeth, well blast drill bits, and crushing plant feed plates.—P. M. C.

Weld-Metal Cutting Edges of Shear Blades for Ingots and Billets. H. Wolff and W. Hummitzsch. (*Schweissen u. Schneiden*, 1952, **4,** Aug., 288–289). An investigation on the build-up by welding of cutting edges of carbon steel shear blades with high-speed steel weld metal is reported.

Design Properties of Brazed Joints for High Temperature Applications. R. L. Peaslee and W. M. Boam. (*Welding J.*, 1952, **31,** Aug., 651–662). A technique for brazing stainless steel has been developed. Dry hydrogen with a dew point

about −40° F. is used as reducing atmosphere in the furnace. Nicrobraz stainless brazing alloy, when used with the appropriate base metal and brazed in a dry hydrogen atmosphere, has satisfactory physical properties. Metals containing large quantities of aluminium or titanium as alloying elements are difficult to braze in a dry hydrogen atmosphere. Illustrations are given of the process as applied to turbine parts.—U. E.

Welding and Brazing of Grey-Iron Castings. H. Spycher. (*Roll'schen Eisenwerke Werkzeit.*, 1951, **22,** Nov., 191–195). Details of methods of welding and brazing cuts and cracks in heavy castings are given, including the gas welding of a crack 18 in. long in the cylinder (4 in. thick) of a hydraulic press.

Brazing with Metal Powder. H. W. Greenwood. (*Machinery*, 1952, **81,** July 17, 98–99). Metal powders may be used in brazing operations on large or small parts composed of steel, cast iron, malleable or wrought iron, or other metals or alloys.—E. C. S.

The Use of Salt Baths for the Soldering of Metals. L. E. Fedotov. (*Stanki i Instrument*, 1951, No. 3, 33). [In Russian]. An account is given of the soldering of metals by immersion in a bath of fused salt. The copper-soldering of flanges on to alloy steel tubing (C 0·2%, Ni <0·30%, Cr < 0·20%) is described. The work, with a suitably arranged copper wire, was totally immersed for 30–60 sec. in a bath of barium chloride at 1120° C. Air being totally excluded, no flux was required. The soldering time must be determined experimentally, but the temperature of the bath should usually be 30–60° C. higher than the melting point of the solder. The gap between the faces should not exceed 0·1 mm.

Jointing of Metals with Metals and with Non-Metallic Materials. K. Meyerhans. (*Metall*, 1952, **6,** May, 229–240). The author describes the use of synthetic resins, notably Araldit and Redux, for joining metals to metals and to ceramics, wood, rubber, and plastics. Joint design and strength are discussed, and illustrations of typical applications are given.—J. G. W.

The Oxygen Torch for Cutting Holes into Concrete, Minerals and Steel. R. Wolf and E. Zorn. (*Schweisstechn.*, 1952, **6,** Feb.–Mar., 13–18). The oxygen torch for cutting purposes is described, and the economic aspects of its use for cutting holes in reinforced concrete are discussed.—U. E.

Development of Flame Cutting Machines from Hand Appliance to Modern Machine Tool. H. von Hofe. (*Schweissen u. Schneiden*, 1952, **4,** Sept., 314–325).

Oxygen-Cutting of Nodular-Graphite Cast Iron. (*Foundry Trade J.*, 1952, **93,** Aug. 14, 186). Nodular cast iron can be cut with oxygen as easily as steel. Reasons for this are given.—R. A. R.

Production Cutting and Flame-Hardening. A. B. Jones. (*Welding J.*, 1952, **31,** June, 469–474). Some examples are given of cutting and flame-hardening in the manufacture of mining machinery and conveying equipment.—U. E.

Stack Cutting of Sheet Steel. C. Lieven. (*New Zealand Eng.*, 1952, **7,** May 15, 203–204). Oxy-acetylene cutting equipment and templates for automatic machines are dealt with, and some hints on cutting procedure are given.—P. M. C.

Metal Removal by Oxygen Processes. E. M. Holub. (*Steel Processing*, 1952, **38,** Aug., 388–390, 399 : *Blast Furn. Steel Plant*, 1952, **40,** Oct., 1197–1199). The author classifies oxygen processes both with and without powder, as follows : cutting, lancing, gouging, scarfing, and flame-washing. Several typical applications, covering each method, are briefly outlined and illustrated.—P. M. C.

Electronic Control of Oxygen Cutting Machine Head. (*Engineering*, 1952, **174,** Aug. 29, 284–285). This article describes a motor-driven tracing head for fitting to oxygen cutting machines, developed by the Metropolitan-Vickers Electrical Co., Ltd., Manchester. The tracing head, which is mounted on one arm of a pantograph, is electronically controlled. The head follows an outline of the shape required, which has been drawn out on paper.—M. D. J. B.

MACHINING AND MACHINABILITY

Electric Spark "Machining". (*Machinery*, 1952, **81,** July 10, 57–61). **"Machining" Hard Metals by Electric Spark.** (*Engineering*, 1952, **174,** July 25, 121–122). This article describes a method and equipment developed by Sparcatron Ltd., Gloucester, for machining hard metals difficult to machine by conventional methods. The machine applies a continuous spark ; the workpiece and electrode are immersed in a liquid dielectric, which flows so as to carry away the disintegrated particles. The surface finish obtained is non-directional ; it consists of a series of very small pits which in many applications is advantageous for retaining lubricants.

Spark Machining: Meets Superalloy Challenge. P. Porterfield. (*Steel*, 1952, **131,** July 7, 92–93). The spark-machining process is discussed (*see* "Electric Spark 'Machining,'" above).

Method X, Electro-Mechanical Machining. M. F. Judkins. (*Canad. Metals*, 1952, **15,** May 56–58). Technical details are given of the spark-machining process (*see* "Electric Spark 'Machining,'" above).

Basic Gear Grinding Methods Compared. D. W. Botstiber. (*Iron Age*, 1952, **170,** Aug. 7, 118–121). The three basic methods of correcting gear tooth form and spacing are form grinding, line grinding, and point grinding. The advantages and limitations of these methods are discussed. Hardened and ground gears can have a transmission capacity 300% greater than that of a conventional heat-treated and cut gear, because of their increased strength and wear resistance.

Coated Abrasives Grind Down Costs. R. E. Hall. (*Steel*, 1952, **131,** Aug. 25, 92–97). A number of examples of the application to metal finishing of coated abrasives are given. Considerable gains in time and lowered cost compared with previous finishing methods are claimed. Contoured surfaces can readily be finished and, if necessary, contoured contact wheels can be used. A simple wheel-tensioning device can convert standard grinding lathes for operation with coated abrasive belts.—A. M. F.

Machinability of Metals. F. W. Boulger. (*Canad. Metals*, 1952, **15,** Feb., 52). The properties of a metal affecting its machinability are given as the shear strength, compressive stress normal to the shear plane effect on the shear strength, and the coefficient of friction between the metal and the cutting tool. Surface finish becomes important when it affects the appearance, fit, or life. Methods of improving it are given.—J. C. B.

Hot Machining Methods for Difficult-to-Machine Metals. A. A. Caminada. (*Mat. Methods*, 1952, **36,** July, 98–100). The author surveys the latest hot-machining techniques and shows that the method can reduce tool wear, increase cutting speed, and improve surface finish. (12 references).—P. M. C.

Electro-Machining of Carbides and Other Hard Compacts. A. H. Allen. (*Metal Progress*, 1952, **62,** Aug., 87–89, 142). A review is made of electrolytic, electro-sparking, electro-arcing, and ultrasonic processes, and combinations thereof under development, which may prove acceptable substitutes for diamond wheel-grinding. Owing to the need for conservation of diamond bort, it is stressed that these processes should be investigated fully and developed rapidly into large-scale industrial methods.—B. G. B.

Making Shaped Electrodes for the Electric-Erosion Machining of Metals. S. I. Veselovskii. (*Stanki i Instrument*, 1951, No. 2, 34). [In Russian]. A very brief account is given of the preparation of shaped electrodes, in which rings of the appropriate shape are turned from a hollow brass cylinder. Each ring is then cut into several segments which are straightened by hammering.—S. K.

Proper Cutting Fluids Can Step Up Machining. R. F. Huber. (*Steel*, 1952, **130,** June 23, 72–76). The material being machined, the type of operation being carried out, and the machine tool material are some of the factors which influence the choice of cutting fluid. The properties of various additives are discussed, and a table of cutting fluid recommendations for use under different conditions is given.

Plunge Cut Grinding. W. J. Constandse. (*Machinery*, 1952, **81,** Sept. 11, 525–527). Plunge cut grinding is a method by which journals and bearing surfaces can be finished to very close tolerances. Machines with two wheel slides, an angular wheel slide grinding machine, and truing devices for grinding wheels are described.—R. A. R.

High-Speed Internal Grinding with Porous Discs. V. P. Nazarov and N. V. Nagarova. (*Stanki i Instrument*, 1951, No. 2, 21–24). [In Russian]. Using grinding wheels with bulk densities approximately 15% less than normal, the surface quality produced improved with increasing grinding rate more sharply than with normal wheels.—S. K.

The Influence of Some Technological Factors on the Quality of a Surface Machined by Grinding. D. M. Tarasenko. (*Stanki i Instrument*, 1951, No. 4, 25–26). [In Russian]. An investigation of the effects of grinding speed, and depth and feed on the quality of the surface produced during the grinding of specimens of alloy and carbon steels is reported.

The quality improved with increasing speed of grinding, for all the steels. Increasing depth of grinding resulted in a coarser surface, the quality being better the harder the steel.

Application of Pneumatic Gauging to the Measurement of Surface Finish. M. Graneek and H. L. Wunsch. (*Machinery*, 1952, **81**, Sept. 25, 701–706 ; *Engineer*, 1952, **194**, Sept. 19, 387–389 ; *Machinist*, 1952, **96**, Oct. 25, 1725–1730). Details are given here of a special design of measuring jet suitable for the examination of both flat and cylindrical surfaces. Results obtained with the pneumatic comparator on a series of ground and turned surfaces having different grades of finish showed a reasonably linear relation with the corresponding centre-line average readings.—E. C. S.

Finishing Methods. L. F. Spencer. (*Machinist*, 1952, **96**, Oct. 18, 1704–1708). The author summarizes the latest practices in finishing stainless steel, including grinding, buffing, and polishing.—E. C. S.

What Are the Rules for Machining Stainless ? G. J. Stevens and V. Bankoski. (*Machinist*, 1952, **96**, Oct. 18, 1694–1703). Rules governing tooling, machine operation and cutting fluids set up by the Rustless Division, Armco Steel Corp. are given.—E. C. S.

High-Speed Turning of Heat-Resisting Steels of the Austenitic Class. A. P. Ivanov. (*Stanki i Instrument*, 1951, No. 2, 3–5). [In Russian]. Difficulties encountered in the high-speed turning of three types of austenitic steel have been investigated. Hard-alloy tipped cutters were used, and the wear of their working surfaces was determined under various conditions. It was established that, at a certain critical feed, in a given range of cutting speed, the ratio of wear at the front and back edges of the cutter changes sharply.—S. K.

Correlation of Machinability with Inclusion Characteristics in Resulphurized Bessemer Steels. L. H. van Vlack. (*Amer. Soc. Met. Preprint* No. 16, 1952). The shape and size of inclusions are affected by the silicon and sulphur contents, the mechanical working, and possibly by the oxygen content of the sulphide phase. The machinability in turn depends on the size or number of inclusions in the steel, when other factors are constant.—E. T. L.

Heat-Treating Boron Steels—Machinability Can be Improved. J. D. Graham. (*Steel*, 1952, **130**, May 12, 94–96). It is shown that most of the problems encountered in machining boron steels can be overcome by conventional heat-treatments, and the hardenability ranges of several types are given. Machining problems mostly arise from substitution of lower carbon contents in the carburizing grades. Medium carbon boron steels quenched and drawn, are no different from the materials they replace, and, in general, the proper heat-treatment of any steel will result in structures that improve machinability.—A. M. F.

Plastic Deformation in the Cutting of Steel. V. V. Kuziushin. (*Stanki i Instrument*, 1951, No. 4, 19–21). [In Russian]. Plastic deformation in the cutting of steel under various cutting conditions is considered, and equations are deduced. Relationships, for a 0·11% carbon steel, between the deformation coefficient and the contraction of the chip for cutting angles ranging from 15° to 115°, and between the coefficients of deformation and contraction on the one hand and cutting speed on the other, are shown graphically. The microstructure of the steel at the point of cutting is shown, together with a chart giving the deformation of each grain.

Uses for Industrial Coated Abrasives Expanded. J. E. Hyler. (*Iron Age*, 1952, **170**, July 24, 98–101 ; July 31, 91–96). The uses to which aluminium oxide and silicon carbide abrasives can be put are described. These are now available on sheets, discs, belts, cylinders, and special shapes besides the usual drum sanders and grinders. Equipment is now available for tool grinding, and centreless grinding and polishing of metal sheet and strip. Machines are also available using formed contact wheels.—A. M. F.

Coated Abrasives Ease Jet Blade Finishing Problems. J. K. McLaughlin. (*Steel*, 1952, **130**, June 9, 90–93). The methods of manufacture of jet turbine blades are briefly described. The use of coated abrasive belts for finishing is advocated. The flexibility of these belts permits grinding or polishing of contoured surfaces and the length of belt dissipates the heat generated by the abrasive action and eliminates the possibility of burning the blade. A number of machines especially built for this process are described.—A. M. F.

Broaching at Ford. (*Canad. Metals*, 1951, **14**, Dec. 33). The difficult operation of broaching thin-walled parts has been eased by a solution developed by the Ford Motor Co. in co-operation with Colonial Tool Co. Ltd., Windsor, Ontario. Side pressure is eliminated by using ' nibbler ' brooches for the inner faces. A 15-ton, 66-in. single-ram machine broaches the universal joint yoke and flange at the same time.—J. C. B.

The " Metalclad " Fusion Cutter. (*Machinery Market*, 1952, Sept. 19, 909). The machine described is designed to cut $1\frac{1}{2}$-in. thick armour plate at 3 in./min. The machine operates on the principle that if a band is run at a high speed and a steady pressure is exerted on the material to be cut, intense heat is generated in front of the cutting edge, which plasticizes the material at this point and so permits rapid cutting.

CLEANING AND PICKLING

Rust Inhibitor Hosed on Large Steel Parts. (*Iron Age*, 1952, **170**, July 31, 98–100). A new cleaner is being used in the Cleveland-Cadillac tank plant. This combines the operations of grease and tar removal with a solvent, cleaning of chips, and producing a rust-inhibitive phosphate coating. No heating of parts, cleaning solution, or rinse, is required, and air-blast drying is used.—A. M. F.

A Portable Surface Cleaning Plant. (*Engineer*, 1952, **194**, Sept. 19, 392). A description is given of a portable shot-blasting plant developed for the preparation of metal surfaces for painting. The plant is manufactured by Vacu-Blast Ltd., London. This machine projects the abrasive material and at the same time picks up the scale, dirt, or dust by a vacuum return system. The abrasive material is then automatically reclaimed.—M. D. J. B.

A Rotary Coil Pickling Plant. *Iron Steel Eng.*, 1952, **29**, Oct., 128–129). This article describes a greatly simplified, low-cost pickling plant in which the coils are loaded into cages and rotated at about 20 r.p.m. in the pickling tank. The equipment, developed in Great Britain, is designed to handle from 5000 to 18,000 tons/annum of commercial grades of coiled hot-rolled strip in widths from 2 to 14 in. and in gauges from 0·042 to 0·188 in.—M. D. J. B.

Protecting Metal Surfaces with Paint. A. G. Gray. (*Metal Progress*, 1952, **62**, Aug., 79–81, 142–146). Methods of preparing surfaces for protective coating are described ; they include grit and shot blasting, wire brushing, flame conditioning, and chemical methods. Among the abrasives now available are malleable and chilled iron shot and grit, copper shot, and high-carbon steel shot. The theory of paint protection is reviewed and the selection of pigments is considered. Tests showed that incorporation of a pigment into phenolic and butyral resin vehicles reduced the resistance of the paint to penetration by salts.—B. G. B.

A Non-Electrolytic Smoothing Treatment for Steel. W. A. Marshall. (*J. Electrodepositors' Tech. Soc.*, 1952, **28**, 27–46 : *Metal Finishing*, 1952, **50**, Nov., 78–83, 85, Dec., 67–69. Aqueous solutions containing oxalic acid, hydrogen peroxide, and sulphuric acid exert a pronounced smoothing action on ferrous alloys, amounting, with mild steel and white-heart malleable cast iron to a considerable degree of polish. A suitable solution contains the following quantities per 100 c.c.: oxalic acid crystals 2·5 g., H_2O_2 1·3 g., and H_2SO_4 0·01 g. The rate of attack on mild steel is approx. $0{\cdot}4 \times 10^{-3}$ in./hr. at room temperature. Steel smoothed in the solution becomes covered with a protective film which markedly inhibits rusting, and this permits considerable delay between etching and the beginning of electrodeposition without impairing adhesion.

Buffing Compounds. H. K. Hunt. (*Metal Finishing*, 1952, **50**, July, 65–66). The chemical and physical properties of buffing compounds are briefly discussed, mention being made of binder content or greasiness, hardness, saponifiability of the binder, and degree of cut.—J. P.

PROTECTIVE COATINGS

Rebuilding Engine Cylinders and Liners. F. M. Burt. (*Metal Finishing*, 1952, **50**, Aug., 67–72). The methods are described by which worn diesel locomotive cylinder liners are cleaned, reground, built up with an electrodeposit of iron if necessary, and finished with a porous chromium deposit. The porosity consists of a series of channels or depressions for retaining lubricant, which are produced by anodic treatment of the chromium.—J. P.

High Operating Efficiency in Job Lot Finishing of Small Parts. F. L. Bonem. (*Products Finishing*, 1952, **16**, Aug., 18–30). The methods employed by Shakeproof Inc., Division

of Illinois Tool Works, for zinc, cadmium, copper and nickel plating, chromating, burnishing, and tinning of pre-assembled steel screws and nuts, stamped gears, lock washers, metal screws, and other fastening devices are described and illustrated.—J. P.

Periodic Reverse Current Electroplating. E. E. Halls. (*Product Finishing*, 1952, **5**, Aug., 61–73). The advantages claimed for periodic reverse current plating are increased rate of deposition, increased density, smoothness and brightness of deposit, possibility of thicker deposits, reduced porosity, and better metal distribution. The procedures by which these properties may be obtained are discussed. with particular reference to copper, zinc, and nickel plating.—J. P.

Research on Electrodeposition. II. W. Blum. (*Indust. Finishing*, 1952, **5**, Aug., 45–48). Some items and results of the research programme of the U.S. National Bureau of Standards are described. These include exposure of plated coatings, non-destructive thickness testing (the magnetic gauge), adhesion, stresses and hardness of coatings, electrodeposition of several elements from non-aqueous solutions, the mechanism of electrodeposition, and throwing power.

Ion Exchange—a New Technique for Metallurgists. B. Ostrof and E. E. Thum. (*Metal Progress*, 1952, **62**, July, 67–74). The use of ion exchange methods for water purification is discussed with special reference to the use of demineralized water in electroplating. The economic recovery of iron, copper, and zinc from pickling wastes is described, and the separation of rare earths by ion exchange is discussed.—B. G. B.

Motor-Generators for Electroplating. E. G. Schroeder. (*Metal Finishing*, 1952, **50**, Sept., 75–77).

A Course in the Principles and Practice of Electroplating. L. Serota. (*Metal Finishing*, 1952, **50**, Aug., 73–76). The syllabus of an evening school technical course is briefly described.—J. P.

Electroplating Steel Springs. (*Metal Ind.*, 1952, **81**, Aug. 29, 167, 168). Recommendations are made by the Coil Spring Federation Research Organization covering the principles which should be observed in the cleaning and electroplating of hardened and tempered steel springs, or of springs made from cold-worked steels, having in view the need for avoiding loss of ductility in the final product. The recommendations aim at avoiding hydrogen embrittlement.—P. M. C.

Protective Finishes—Phosphate Coatings for Military Equipment. (*Machine Design*, 1952, **24**, Jan. 108–114, 196–198 ; Feb., 141–145, 222). The principle of chemical conversion coatings is that the surface of a metal is changed by chemical reaction with a natural or artificial environment to an insoluble phosphate, oxide, or other salt which remains as an integrated coating. These coatings may be applied for : (1) Colouring, (2) corrosion-proofing ; (3) improvement of wear resistance, safe ' breaking-in ' of friction or rubbing surfaces ; or (4) improving paint adhesion. Part I deals with finishes and undercoating, and Part II with rustproofing and protecting friction surfaces.—E. C. S.

Automatic Acid Bright Dip Has Many Advantages. K. W. Bennett. (*Iron Age*, 1952, **170**, July 24, 108–110). A fully automatic bright dip machine with a capacity of 1 ton/hr. is described. The process includes a double cleaning, rinse, chromate treatment, cold rinse, and hot rinse. The total length of the line is only 37½ ft. and it is operated by two men. The 'dwell' period at any station can be set to a fraction of a second.—A. N. F.

The Ward Leonard Chromaster Hard Chrome Plating Unit. (*Machinery*, 1952, **81**, Aug. 21, 348–349). A demonstration was recently given by Refrigerator Component Ltd., London, of the Ward Leonard Chromaster industrial hard chromium plating unit. The unit demonstrated was the type A–50 and its operation is described. The chromaster units are primarily designed for the treatment of small tools such as taps, drills, cutting tools, dies, and gauge and similar components.—E. C. S.

Some Unusual Chromium-Plating Operations. A. W. Brown. (*Machinery*, 1952, **81**, Aug. 28, 362–366). To facilitate experimental and research work on chromium plating problems the Cro-Plate Co., Inc., Hartford, Conn., has set up an engineering service division. Some problems dealt with, including the plating of rifle breeches, are described.

Knurling Steps up Wear Resistance of Chrome Plate. E. F. Discoe. (*Machinist*, 1952, **96**, Nov. 1, 1781).

Electroplating Large Complex Shapes. F. K. Savage, A. K. Graham and E. P. Strothman. (*Mat. Methods*, 1952, **36**, Aug., 94–97). The successful use of electroplated coatings of nickel on large chemical equipment has resulted in great savings on the cost of clad metals and corrosion-resistant alloys. The plating methods, cleaning and preparing cycles, and unusual plant involved, are described. Items 600 sq. ft. in area have been successfully plated in a single operation.

Electricity Speeds Production of Tinplate from Trostre. (*Electrician*, 1952, **149**, Sept. 19, 829–832). **Trostre and Tinplate.** (*Electrical Times*, 1952, **122**, Oct. 9, 637–643). **The Trostre Tinplate Works.** (*Engineer*, 1952, **194**, Sept. 26, 413–414 ; Oct. 3, 447–449 ; Oct. 10, 481–484). **The Trostre Tinplate Works.** (*Metal Bulletin*, 1952, Sept. 20, 16–22, 27). **Production of Tinplate: The New Works of the Steel Company of Wales, Limited, at Trostre.** (*Iron Coal Trades Rev.*, 1952, **165**, Sept. 19, 647–649). **The Trostre Tinplate Works of the Steel Company of Wales.** (*Engineering*, 1952, **174**, Sept. 19, 382). A description is given of the new Trostre tinplate works of the Steel Company of Wales Ltd. The different operations carried out in the works in the production of tinplate are outlined.—G. F.

A New Process for Electro Deposition of Aluminium. (*Machinery*, 1952, **81**, Aug. 14, 284–285). A process has been developed, by D. E. Couch and A. Brenner of the U.S. National Bureau of Standards, in which ductile deposits of the metal are obtained from a new type of organic plating bath. This process is expected to have important applications for electroforming parts, such as wave guides to close tolerances, and for providing various types of equipment with thin protective coatings of aluminium. The bath is prepared by adding either lithium hydride or lithium aluminium hydride to an ethyl ether solution of anhydrous aluminium chloride. For the best results, the ether should be anhydrous and alcohol-free.—E. C. S.

Developments in the Al-Fin Process. (*Machinery*, 1952, **80**, June 12, 1024–1027). Some details of applications of the Al-Fin process of chemically bonding aluminium and its alloys to cast iron and steel are given. In England the process is operated by Wellworthy Piston Rings Ltd., It is not applicable to finished parts since the bi-metallic component is produced by casting the aluminium about the iron or steel insert after the latter has been subjected to immersion in molten alloy to provide the bonding surface.—E. C. S.

Impregnating Mild Steel. (*Brit. Eng.*, 1952, **35**, Sept., 100–101). The Penetral process, developed by Follsain-Wycliffe Foundries, Ltd., can render mild steel resistant to high-temperature oxidation, thereby saving alloy steels. Up to $\frac{1}{16}$ in. is impregnated to form a complex Al–Fe alloy containing chromium and silicon. Resistance to oxidation is obtained up to 1000° C., and protection against dry sulphurous gases is also given. Ni–Cr steels may also be protected in the same way. The process is applicable to heat-treatment plant used in the engineering industries, cyanide pots, carburizing equipment, and pyrometer sheaths.—K. E. J.

From Researches on Coating Steel Sheets. F. Staub and J. Szram. (*Hutnik* (Katowice), 1952, **19**, 6, 202–205). [In Polish]. Methods of coating metals are outlined. Coating steel with a Cu–Zn alloy by rolling was investigated. Operating conditions under which optimum properties of the product were obtained are given.—V. G.

Some Factors in Spray-Silvering. P. B. G. Upton, G. W. Soundy, and G. E. Busby. (*J. Electrodepositors' Tech. Soc.*, 1952, **28**, 103–113).

Rhodium Plating. F. Spicer. (*Sheet Metal Ind.*, 1952, **29**, May, 462–464). Rhodium may be deposited fairly easily up to 0·0005 in. in thickness. The main solutions are sulphates, phosphates, and combinations of the two. Compositions and methods of operating these solutions are given. Steel, iron, and articles containing lead solder require a flash coating of copper or silver before rhodium plating.—P. M. C.

On the Formation of Spots on Electrodeposited Noble Metals. W. Burkart. (*Metalloberfläche*, 1951, **5**, May, B65–B66). It is shown that spots appearing on silver plating arise through the action of local galvanic couples located at pores in the surface of the base metal. Spot formation can be avoided by having a high polish on the surface to be plated and by increasing the thickness of the silver deposit so that diffusion of oxygen through the silver to pores in the base metal cannot take place.—P. F.

Finishing of Harley-Davidson Motorcycles. (*Indust. Heating*, 1952, **19**, May, 885–890). Three separate paint finishing systems complete with wash, dry off, and paint-baking ovens which are used, are described in detail.—B. G. B.

The Finishing of Motor-Car Bodies. A. A. B. Harvey. (Sheet and Strip Metal Users' Tech. Assoc. : *Sheet Metal Ind.*, 1952, **29,** May, 451–458). The plant and processes used by Briggs Motor Bodies Ltd., for cleaning, phosphating, priming, and finishing, are described.—P. M. C.

Plates and Sections : Phosphating Treatment at Appleby-Frodingham. (*Iron Steel*, 1952, **25,** June, 282). A plant has recently been installed at the Appleby-Frodingham Steel Co., for the pickling, phosphatizing, and priming of plates and sections by the Footner process. This article describes the plant and operations, and outlines the advantages claimed for this process.—G. F.

Phosphoric Acid Treatment Improves Paint Adherence, Corrosion Resistance. W. G. Patton. (*Iron Age*, 1952, **170,** July 17, 138–139). The Nielizing process of pre-paint treatment is briefly described. It can be used for both uncoated and zinc-coated carbon steels and the installation requires little floor space. A phosphoric acid surface conditioner cleans the steel parts at 180° F. in 2–3 min. and this is followed by a hot-water rinse for zinc coated steel ; uncoated steel parts receive an additional dip in a firing solution to neutralize the metal surface.—A. M. F.

Protective Film Formation with Phosphate Glasses. G. B. Hatch. (*Indust. Eng. Chem.*, 1952, **44,** Aug., 1775–1780). Inhibition by glassy phosphates of galvanic corrosion of steel coupled to cathodic metals is mainly due to polarization by an electrodeposited film. A similar process occurs with uncoupled steel : the film appears on local cathodic areas. In differential aeration cells with steel electrodes, the glassy phosphate film also lowers the p.d. between the electrodes, and the inhibitor, and leads to reduction of current flow. (13 references).—K. E. J.

Inhibition of Galvanic Attack of Steel with Phosphate Glasses. G. B. Hatch. (*Indust. Eng. Chem.*, 1952, **44,** Aug., 1780–1786). Glassy phosphate has been found to be a very effective inhibitor for corrosion of steel coupled to copper, and for zinc coupled to either copper or steel ; with steel coupled to copper, this is largely due to increase of cathodic polarization resulting from a phosphate film on the copper. A very similar mechanism exists for phosphate inhibition of attack on steel alone.

Wash Primer Helps Industry Fight Corrosion. R. A. Garling. (*Iron Age*, 1952, **170,** July 24, 102–107). The use of Vinylite resin-based prime coats is shown to combine the most desirable properties of a phosphate metal preparation and zinc chromate primer. It wets the metal surface and has excellent adherence. It is applied by conventional painting methods and dries extremely fast.—A. M. F.

Accelerated Phosphate Coating Fits High Speed Finish Setups. N. Kuperschmid. (*Steel*, 1952, **130,** May 12, 92–93). An accelerated phosphate prepainting treatment is described. The process utilizes a compound called " Anchorite 100 " which changes the surface of the metal into a combination of metallic phosphates. This microcrystalline surface is highly corrosion resistant and holds paint tenaciously to metal.

Vitreous Enamelling. S. Hallsworth. (*J. Inst. Prod. Eng.*, 1952, **31,** July, 313–323). A description is given of the various stages in the vitreous enamelling of metals on a commercial scale.—B. G. B.

Selection of Metals Suitable for Heat Resistant Ceramic Coatings. (*Indust. Heating*, 1952, **19,** Aug., 1500). A recent guide to the temperatures at which metals with ceramic coatings con be used in industrial installations is reviewed. B. G. B.

Factors Affecting the Rusting and Spalling of Titania Enamels Under Alternate Freezing and Thawing in the Presence of Moisture. R. F. Patrick. (*Bull. Amer. Ceram. Soc.*, 1952, **31,** Sept., 334–339). The development of rusting and spalling defects in enamels used in cold-wall refrigerator liners is discussed, and the apparatus used to make accelerated freezing and thawing tests is described. Data concernin the effect of such factors as mill additions, firing conditions, cover-coat and ground-coat frits, on the rusting and spalling resistance are given.—E. C. S.

Fundamental Aspects in Research on Titania–Opacified Enamels. S. S. Cole. (*J. Amer. Ceram. Soc.*, 1952, **35,** July, 181–187).

Comparison of Warpage, Cross-Bend, Torsion, and Expansiometer Ring Tests on Titania Enamels. J. F. Lochridge and G. E. Miller. (*J. Amer. Ceram. Soc.*, 1952, **35,** Apr., 103–106).

A Contribution to the Theory of Enamel Adherence. M. Berg and M. Humenik, jun. (*Bull. Amer. Ceram. Soc.*, 1952, **31,** Sept., 329–331). By quantitative determination of the work of adhesion it was found that : (1) The addition of cobalt oxide in an enamel glass does not increase the glass–iron chemical adherence on an oxide-free iron surface. (2) In the presence of iron ions, cobalt ions, and hydrogen, a concentrated metallic precipitate is formed just above the glass–iron interface. (3) The formation of the metallic phase adjacent to the glass–iron interface probably promotes adherence by relieving the interfacial stresses arising from cooling.—E. C. S.

Radioisotope Study of Porcelain-Enamel Adherence. (*Ceramics*, 1952, **4,** Aug., 37–40). Cobalt oxide promotes the adherence of porcelain enamel to iron and for this reason has long been added to enamel ground coats. By adding radioactive cobalt as a tracer to the frit used in coating, the U.S. National Bureau of Standards has recently obtained new information on the mechanism by which cobalt oxide promotes adherence.—E. C. S.

Ceramics Make Furnace Fixtures Last Longer. D. K. Krosch. (*Machinist*, 1952, **96,** Sept. 6, 1379). Temperature-resistant ceramic coatings have been developed by the Solar Aircraft Co., San Diego, which greatly prolong the life of furnace fittings.—R. A. R.

Some Applications of the Fluorescent X-Ray Spectrometer in Ceramics. R. F. Patrick. (*J. Amer. Ceram. Soc.*, 1952, **35,** Aug., 189–193). The X-ray spectrometer can be used for the determination of specific elements, above titanium (22) in atomic number, in virtually any raw material *e.g.*, iron in pyrophyllite, and niobium in titania. The technique has also been used to determine quantitatively the nickel plated out in nickel pickling, and the amounts of cobalt and nickel deposited on the base metal during the firing of enamel ground coats. The advantages and disadvantages of the technique are discussed.—E. C. S.

Relation of Bubble Structure to Freeze–Thaw Spalling in Porcelain Enamel Cover Coats. G. B. Hughes, H. D. Bowsher, and J. A. Schiefferle. (*Bull. Amer. Ceram. Soc.*, 1952, **31,** June, 199–203). The bubble structure of several enamels was microscopically studied after progressive etching with 24% HF. The data showed a definite correlation between spalling and the number of bubbles per square inch with an apparent diameter of 0·003 in. and larger. The method developed allows the evaluation of relative freeze–thaw spall resistance of enamel cover coats in less than an hour.—E. C. S.

Use of Alkaline Silicates as Binding Agents in Industry, Particularly for Ceramics and Enamels. A. Debecq. (*Silicates Indust.*, 1952, **17,** May, 157–161).

The Increasing-Load Wire Abrasion Tester. H. T. McLean. (*Wire and Wire Products*, 1952, **27,** Sept., 871–873, 928–929). An instrument is described for testing the abrasion resistance of enamel coatings on magnet wire.—J. G. W.

Methods for Testing for Enamel Coating Discontinuities. S. C. Orr. (*Non Destructive Test.*, 1952, **10,** Spring, 23–27). The author outlines the main methods used for revealing discontinuities in enamel coatings. These are the resistance test, the electrolytic test, the transformer test, the high-voltage test, and the use of electrostatically charged powder particulars. The last test was developed by the Magnaflux Corp. and is named Statiflux.—P. M. C.

Manufacture and Application of Porcelain Enamels. L. A. Buckley. (Inst. Australian Foundrymen : *Australian Found. Trade J.*, 1952, **3,** Mar., 9, 11–15). The author deals first with the nature of porcelain enamels and the methods of manufacture, and continues with an account of the wet and dry processes for their application to iron.—P. M. C.

Process Control in Vitreous Enamelling. H. W. Clewes. (*Sheet Metal Ind.*, 1952, **29,** June, 549–556 : *Foundry Trade J.*, 1952, **93,** Sept. 4, 275–281).

Some Factors Involved in the Enamelling of Hollow-Ware. K. H. Broadfield. (Inst. Vitreous Enamellers : *Sheet Metal Ind.*, 1952, **29,** May, 459–461). This is a brief review of the processes involved which include removal of lubricants, pickling, assembly of integral parts, choice of enamels, application, and fusing.—P. M. C.

Drying and Finishing Ovens. J. D. Russell. (*Indust. Heating*, 1952, **19,** Jan., 107–118, 121 ; Feb., 297–306 ; Mar., 495–502 ; Apr. 687–692, 699–700). The horizontal wicket-type oven for the mass-production baking of flat sheet stock is described. Multiple zones control the baking cycle and can be regulated to follow any preselected heating and cooling curve. The tip-up oven and vertical oven are also discussed. The importance of effective air recirculation is considered, and

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an example is cited of an oven with air recirculation in which a coil of wire was brought to 340° F. in 65 min. against 3 hr. without recirculated air.—B. G. B.

Possibilities in Porcelain Enamel. E. Mackasek. (*Steel*, 1952, **130**, Apr. 7, 116–117). It is suggested that the high temperature and corrosion resistance of porcelain enamel coatings can be used in the protection of non-strategic steels to replace non-corrosive metals. Coatings are used extensively for jet motor and rocket components at high temperatures.—A. M. F.

Infra-Red Oven for Baking Enamel Finishes. S. M. Mellor. (*Electrical Engineer and Merchandiser*, 1952, **29**, Apr., 10–12). The selection and use of infra-red heating is reviewed. An installation for baking enamel on steel refrigerator cabinets is described, and details are given of the lamp arrangement and timing control.—B. G. B.

Apparatus for the Determination of the Adhesion of Varnishes to the Base. O. Cuzzer, C. Pizzuto, and L. Caccavo. (*Rev. Mét.*, 1952, **49**, June, 423–435). The apparatus has a loaded knife of variable inclination and is sheathed in a metal softer than the sample studied. The adhesion of marine chrome and marine grey varnishes to various metallic surfaces has been investigated, the former giving the better results.—A. G.

Study on the Electrostatic Spray-Coating Process. S. Yoshida and K. Tamura. (*Report of the Government Mech. Lab.* No. 10, 1952, **3**, 35). [In Japanese]. The following results of this process were obtained : (1) The paint utilization factor is increased ; (2) higher voltage and thinner discharging electrode improve the paint film ; (3) quick-drying solvents are inadequate ; (4) hardness, adherence, and smoothness are unimpaired and (5) uniformity of thickness is satisfactory with comparatively simple shapes. The process was applicable to a large range of articles. A study of a spray gun is discussed, but a final form is not recommended.—J. C. B.

POWDER METALLURGY

Progress in Powder Metallurgy, with Special Reference to High Temperature and Hard Metal Alloys. W. D. Jones. (*J. Chem. Met. Min. Soc. S. Africa*, 1952, **52**, Mar., 216–222). The author discusses such aspects as mixing, milling, pressing, sintering, and shaping in general terms. Particular attention is paid to the manufacture of carbide-tipped tools and high-temperature alloys.—D. H.

Chrome Carbide Provides High Corrosion Resistance. J. D. Kennedy. (*Steel*, 1952, **131**, Aug. 4, 92–94). A new sintered carbide composed of 83% Cr, 2% W and 15% Ni, is described. It is lighter in weight than tungsten carbide, provides extremely high corrosion resistance, has a coefficient of thermal expansion about the same as steel, is completely non-magnetic, and provides high resistance to temperature oxidation. The material can be economically machined to required shapes in the presintered condition.—A. M. F.

Powder Metallurgy Parts in High Temperature Applications. J. J. Harwood. (*Mat. Methods*, 1952, **36**, Aug., 87–91). The author discusses the development of copper infiltrated iron, titanium alloy powders, sintered aluminium powder, molybdenum-base alloys, and ' cermets '. The uses of such new materials for high-temperature service in jet compressors and turbines are reviewed.—P. M. C.

Crushing Strength Requirements Met by Powder Metallurgy. (*Precicion Met. Mold.*, 1952, **10**, Aug., 29). Details are given of the production of a chain roll used in textile machinery. Rolls now made from powdered iron parts have almost the same wear resistance and crushing strength as a fully machined and hardened steel roll.—D. H.

Hot Molding Increases Scope of Powder Metallurgy. J. F. Kuzmick. (*Machinist*, 1952, **96**, Aug. 30, 1352–1356). The equipment used, some of the applications, and the advantages and limitations of the process are described.—E. C. S.

Production, Properties, and Applications of Sintered Hard Metals. A. Merz. (*Technik*, 1952, **7**, June, 295–299).

Sintered Materials for Industry, Especially the High-Melting-Point Metals and Alloys. K. Pschera. (*Technik*, 1942, **7**, June, 299–303).

PROPERTIES AND TESTS

Metal Testing at Ford Labs. H. Acker. (*Canad. Metals*, 1952, **15**, Jan., 34–38). A review of the four laboratories of the Ford Motor Co. of Canada is given. The main laboratory is devoted to physical and chemical testing of metals, and also other tests on miscellaneous materials going into a Ford car. Quality control is used to control variables in the manufacturing processes. Some process research is also carried out.

A Survey of " Statistical Effects " in the Field of Material Failure. W. Weibull. (*Appl. Mech. Rev.*, 1952, **5**, Nov., 449–451). Any simple test, repeated sufficiently often with nominally identical specimens, shows a scatter of the ultimate strength caused by many unknown factors. The strength is thus a statistical quantity which has to be defined, not by a single figure but by a statistical distribution function. Many ' effects ' which are incompatible with the classical theory may be explained by this concept ; they are called ' statistical effects.' Size, shape, stress-distribution, stress-type and notch effects are considered. (63 references).—B. G. B.

The Development of Engineering Metrology. F. H. Rolt. (*Inst. Prod. Eng. Sir Alfred Herbert Paper*, 1952). A survey of the development of engineering metrology during this century is presented. The development and manufacture of slip gauges, engineering standards of angles, and machines for measuring and inspecting screw gauges at the National Physical Laboratory are described. A number of measuring instruments developed by Carl Zeiss and the Société Genevoise d'Instruments de Physique are described. Recent examples of improved methods of internal measurement are discussed. Examples of proposed methods of measuring surface finish are reviewed. (150 references.)—B. G. B.

Calibration of Load Measuring Devices. B. L. Wilson. (*N.P.L. Symposium on Recent Developments and Techniques in the Maintenance of Standards*, May, 1951, 1–6 : H.M. Stationery Office, 1952). The performance and use of the dead-weight machines now available at the National Bureau of Standards are discussed. The machines serve as the primary standard for the calibration of portable elastic calibration devices used for calibrating machines for testing materials.—B. G. B.

The Primary Load Standard. F. Aughtie. (*N.P.L. Symposium on Recent Developments and Techniques in the Maintenance of Standards*, May, 1951, 7–11 : H.M. Stationery Office, 1952). The 50-ton dead-weight standard constructed at the National Physical Laboratory in 1944 is described. For small loads a 5-ton standard has since been developed.—B. G. B.

The Present State of a Lightwave Standard of Length. E. Engelhard. (*N.P.L. Symposium of Recent Developments and Techniques in the Maintenance of Standards*, May, 1951, 13–23: H.M. Stationery Office, 1952). Recent developments of optical methods for the accurate measurement of the standard of length are reviewed. The Kosters double-prism method is described in detail. The need for agreement upon which spectral line shall serve as the base for measurements is stressed.—B. G. B.

The Interferometry of Length. H. Barrell. (*N.P.L. Symposium on Recent Developments and Techniques in the Maintenance of Standards*, May, 1951, 23–34, H.M. Stationery Office, 1952). A number of interferometric methods of measuring length are described. Several monochromatic sources of radiation which have been proposed as standards are discussed. At present the choice would appear to be between the green line of $^{198}_{80}$ Hg and the yellow-green line of $^{86}_{36}$ Kr.

The Establishment of the Photometric Scale at the National Physical Laboratory. J. W. T. Walsh. (*N.P.L. Symposium on Recent Developments and Techniques in the Maintenance of Standards*, May, 1951, 63–68 : H.M. Stationery Office, 1952). The use of the primary standard of light (black body radiation at the temperature of solidification of molten platinum) and the preparation of various substandards from this are described.

Co-ordination of the Photometric Units at the Bureau International des Poids et Mesures. J. Terrien. (*N.P.L. Symposium on Recent Developments and Techniques in the Maintenance of Standards*, May, 1951, 68–74 : H.M. Stationery Office, 1952). An account is given of the activities of the above Bureau and of the international collaboration on photometric standards which it sponsors.—B. G. B.

The Measurement of Time, Frequency and Velocity. L. Essen. (*N.P.L. Symposium on Recent Developments and Techniques in the Maintenance of Standards*, May, 1951, 88–100 : H.M. Stationery Office, 1952). The precision of present physical measurements has increased to such an extent that the limitations of the present fundamental standards in respect to their difficulty of access and inaccuracy of use are now beginning to be felt, and possible alternatives must be seriously considered. A number of possible alternatives to these fundamental standards are discussed.—B. G. B.

On Discontinuous Plastic States with Special Reference to Localized Necking in Thin Sheets. R. Hill. (*J. Mech. Phys. Solids*, 1952, **1,** Oct., 19–30). Discontinuities of stress, velocity, and surface slope, which may occur in a plastic-rigid sheet deformed in its plane, are investigated. This leads to an examination of the modes of necking in notched strips in the tensile test ; the constraint factors at the yieldpoint are calculated for notches with sharp and curved roots.—J. G. W.

Cold Rolled Fillets. R. J. Love. (*Engineering*, 1952, **174,** Aug. 8, 164). The author describes a process devised at the Motor Industry Res. Assoc. for strengthening fillets ; it consists of cold-rolling with steel balls. The method has given striking result on cast crankshafts. Three balls, equally spaced, are rolled round the fillet ; they are held in place by a loading ring with an internal 45° chamfer. The fillets are much superior to those produced by conventional methods, since, apart from the beneficial cold-working effect, they are very smooth, truly circular and of closely controlled radius.—M. D. J. B.

The Yield Phenomenon in Polycrystalline Mild Steel. W. M. Lomer. (*J. Mech. Phys. Solids*, 1952, **1,** Oct., 64–73). The author summarizes the most important macroscopic features of the yield-point in polycrystalline iron, and singles out the fundamental properties which must be explained by a theory for it to be acceptable. He suggests that the two essential quantities to be derived from any theory are the magnitude of the Lüders strains, and the degree to which they are locals.

Plastic Instability under Plane Stress. H. W. Swift. (*J. Mech. Phys. Solids*, 1952, **1,** Oct., 1–18). The degree of thinning which sheet metal may undergo in deepdrawing, stretch-forming, and the Erichsen, Guillery, and other bulging tests, is limited by the onset of local necking under diminishing load. The conditions for instability under plane (biaxial) stress are examined theoretically for a solid which yields according to the Mises-Hencky condition, and hardens according to a prescribed relation between the root-mean-square value of shear stress and of strain increment. Theory and experiment are compared for the hydrostatic bulge test, and cylindrical shell pressing is briefly discussed.—J. G. W.

Note on the Strain Aging of Iron Single Crystals. A. N. Holden and F. W. Kunz. (*J. Appl. Phys.*, 1952, **23,** July, 799). Tensile tests were performed on annealed single crystals of high-purity iron containing 0·005% carbon. They had either very small initial yield points or no initial yield point at all, but, after subjecting the crystal to a tensile stress of roughly two-thirds of the yield value (of stress) for 1 hr. before testing, a marked increase in the upper yield point occurred over that of an unstressed crystal. When a crystal was deformed in tension and aged while still subjected to a high stress, a higher upper yield stress was observed in a subsequent test than was observed when the deformed crystal was aged with the external stress relaxed.—R. A. R.

The Effect of Water Corrosion and Shot Peening on Fatigue Strength of Mining Drill Rods. T. W. Wlodek. (*Trans. Canad. Inst. Min. Met.*, 1952, **55,** 292–300 ; *Canad. Min. Met. Bull.*, 1952, **45,** Aug., 470–478). The results of research into the fatigue properties of a 0·80% carbon steel and Ni-Cr-Mo steel for mining drill rods are analysed and described in detail. The effect of water corrosion on the fatigue strength of drill rods was investigated, and also the effect of shot-peening. Eight stress/cycle curves for as-rolled and shot-peened drill rods under both dry and water-corrosion conditions are derived and discussed.—T. E. D.

Spring Failures and Their Causes. F. P. Zimmerli. (*Metal Progress*, 1952, **62,** July, 84–88). Formulæ for determining stress in extension or compression springs, torsion springs, and flat cantilever springs, and the maximum values for infinite life are given. It has been found that the endurance limit of valve springs is independent of the steel analysis. The surface and subsurface defects considered include hardening cracks, seams, and damage from pits, die marks, and tool marks. Corrosion caused by condensation of the products of combustion is a common cause of breakage. Other defects mentioned are improper heat-treatment and setting.—B. G. B.

Residual Stresses in Springs. O. G. Meyers. (*Product Eng.*, 1952, **23,** July, 148–149). Longer life and increases up to 30% in the maximum allowable working stress can be obtained if residual stresses are controlled. Stress relief and cold setting are the two controlling factors and the effects of these are briefly discussed.—A. M. F.

The Fatigue of Metals. M. Roš. (*Rev. Mét.*, 1952, **49,** July, 497–510). The determining factors and theories to be considered in static and fatigue failure are outlined for (*a*) homogeneous stress distribution without notching, and (*b*) heterogeneous stresses with or without notching. The Coulomb Mohr hypothesis is applicable to unnotched specimens in all conditions and to notched specimens under heterogeneous stress. The limiting states under triaxial stress for both static and fatigue rupture are related to those under biaxial stress, and methods of procedure for determining these limiting values are illustrated graphically for many stress systems.—A. G.

Fatigue of Metals. J. A. Pope. (*J. Inst. Prod. Eng.*, 1952, **31,** Sept., 378–400). A review is presented of some of the theories relating to the fatigue of metals. If expense is no object, and the special steel necessary is obtainable, then nitriding is the best method of improving the fatigue properties. If, however, ordinary steels must be used and the process must be an economic proposition, then shot peening deserves the greatest attention by research workers and industry. If components are required in the machined condition then 0·002 in. may be removed from the surface after shot peening ; this final machining in most cases actually improves the fatigue strength still further.—B. G. B.

A Critical Look at Fatigue Equations. E. H. Schnette. (*Product Eng.*, 1952, **23,** July, 150–151). It is pointed out that empirical fatigue equations are unreliable and often misleading. This is demonstrated by the application of four such equations to test data for a magnesium alloy. Values of the endurance limit thus found range from 0 to 19,000 lb./sq. in.—A. M. F.

The Effect of Shot Peening on the Fatigue Life of Steel. A. G. H. Coombs. (British Assoc. : *Iron Coal Trades Rev.*, 1952, **165,** Oct. 3, 759 : *Engineering*, 1952, **174,** Oct. 24, 545–546 ; Oct. 31, 580–581). This article summarizes the results of an examination of the effect of shot velocity, size, and energy on the penetration of the shot-peening effect, with special reference to the fatigue life. It is shown that considerable improvement in fatigue life is obtained by polishing sufficient material from peened specimens to bring into their surfaces the layer in which maximum fatigue life was obtained.

A Lock Nut and Nut Fatigue Testing Machine. (*Engineer*, 1952, **194,** Aug. 29, 284–285). Details are given of a new lock nut and vibration testing machine for nut and bolt assemblies. The lock nut and nut fatigue testing machine is manufactured by the Lester Lock Nut and Washer Co., Ltd.—M. D. J. B.

Proof Testing of Full Sized Automobile Components. A. L. Boegehold. (*Metal Progress*, 1952, **62,** Aug., 82–86). The determination of endurance limits of rear-axle components is described. The testing machine used has a motor-driven propeller shaft and power is taken off by electric generators at each end of the axle. A special apparatus for testing crankshafts, a constant-load fatigue tester, and a hydraulic-driven fatigue tester capable of imposing very high loads at small amplitudes are among the other machines described.

Hardness of Various Steels at Elevated Temperatures. F. Garofalo, P. R. Malenock, and G. V. Smith. (*Amer. Soc. Met. Preprint No.* 18, 1952). A hot-hardness tester of the static loading type, in which a number of steels have been tested up to 1500° F. is described. The results suggest a relationship between hot-hardness and creep-rupture strength. A similar relation is observed between hot-hardness and the ultimate tensile strength in short-time hot-tensile tests.

Temperature Dependence of the Hardness of Pure Metals. J. H. Westerbrook. (*Amer. Soc. Met. Preprint No.* 17, 1952). A review of previous literature shows that the hardness (H) of pure metals is well represented by a function $H = Ae^{-BT}$ where the constants A, B have one set of values at low temperatures where slip is the primary deformation mechanism, and another one corresponding to viscous flow. The transition is at a temperature $0{\cdot}55xT_{mp}$,2 where T_{mp} is the absolute melting point of the metal. This transition temperature is varied slightly by impurity, and alterations in grain size and strain rate. The values of A increase with decreasing symmetry in the crystal structure at low temperatures. The constant B is associated with the rate of change of heat content with increasing temperature. Because of these regularities, it is possible to calculate approximately the hardness of pure metals as a function of temperature.—E. T. L.

On the Concept of Homogeneity. H. Franssen. (*Metall*, 1952, **6**, June, 302–305). The point that homogeneity of a substance is a relative concept, depending on the quantity measured, is made deductively and illustrated by Vickers and Brinell hardness measurements.—J. G. W.

An End-Quench Test for Determining the Hardenability of Carburized Steels. F. X. Kayser, R. F. Thomson, and A. L. Boegehold. (*Amer. Soc. Met. Preprint No.* 13, 1952. A new shape of test-bar for case-treated steels is described. The cross-section is somewhat trapezoidal, and only the non-parallel sides are case-treated, with hardness measurements being taken from the parallel sides. Hardness impressions can be made at any point in the case or in the core without the necessity for grinding, and they are made perpendicular to the carbon concentration gradient. Furthermore, the results obtained are directly comparable with those obtained by the normal Jominy test. The results on seven steels indicated that the hardenability of boron steels goes through a maximum at about 0·7% carbon.—E. T. L.

Prediction of End-Quench Hardness in Steels Simplified. M. A. Orehoski and J. M. Hodge. (*Iron Age*, 1952, **170**, July 17, 125–129). A method is developed by the authors for estimating the Jominy curve from the first $\frac{1}{16}$ in. to the region of 50% martensite if both carbon content and ideal diameter are known. Accuracy of the method is reasonably good, provided that the two basic factors are reliable. The method is not too accurate on boron steels.—A. M. F.

Temper-Brittleness: A Critical Review of the Literature. B. C. Woodfine. (*J. Iron Steel Inst.*, 1953, **173**, Mar., 229–240). [This issue].

Some Aspects of Temper-Brittleness. B. C. Woodfine. (*J. Iron Steel Inst.*, 1953, **173**, Mar., 240–255). [This issue].

Accelerated Strain-Aging of Commercial Sheet Steels. L. R. Shoenberger and E. J. Paliwoda. (*Amer. Soc. Met. Preprint No.* 20, 1952). Variations in the ageing characteristics of normal rimming steels are attributed to the effect of ferrite grain size and skin rolling. An ageing test was developed using heat-treated samples and employing time–temperature relationships to obtain varying degrees of accelerated ageing. These relationships were correlated with room temperature ageing, so that accurate predictions could be made of the age-hardening of freshly skin-pass-rolled sheet and strip steels.

The Problem of Natural and Artificial Ageing. R. Mitsche and A. Legat. (*Betrieb u. Fertigung*, 1952, **6**, June, 105–108). The tendency of deep-drawing quality steel strip to natural ageing, as measured by changes in the Erichsen test value, cannot be judged by its behaviour under artificial ageing at 250° C. It is also shown that some steels suffer a serious reduction in the Erichsen value by being cold rolled with a reduction of as little as 3%, without any subsequent ageing.

Hydrogen Embrittlement and Hairline Cracks in Steels. A. B. Chatterjea and B. R. Nijhawan. (*J. Sci. Indust. Res.*, 1952, **11**, Apr., 158–161). The literature is reviewed. (45 references).—E. T. L.

Machine for Wear-Testing Roll Materials. A. L. Chestnov. (*Stanki i Instrument*, 1951, No. 4, 23–25). [In Russian]. Objections to the machines at present used for the wear-testing of roll materials are listed and a new machine for testing shafts up to $3\frac{1}{4}$ in. in dia. is described; it is based on the principle of the wearing-away of the measuring instrument in the course of the measurement. A rigid claw is in contact with and moves over the rotating shaft, and its displacements resulting from this movement and from wear are automatically recorded. When all the surface of the shaft has been used, it can be re-turned and used for further tests.—S. K.

Developments in the Study of Metal and Plastic Slides. (*Production Engineering Research Assoc.*, 1952, Memorandum). Tests by the above association on cast iron and laminated plastic slideways are described. Friction properties of some plastics, and coefficients of friction of unlubricated metals and plastics, are given. The oil film between sliding surfaces has also been studied.—T. E. D.

Theory of Mechanical Wear. J. F. Archard. (*Research*, 1952, **5**, Aug., 395–396). An attempt is made to provide a simple quantitative theory of wear based on the small real area of contact between rubbing surfaces. Expressions are derived which illustrate ' models ' of wear in terms of lump and layer removal. Experimental results showing wear rate proportional to the load are more readily understood if the real area of contact is formed by plastic deformation of the contacting asperities and the material is removed in lumps, than if wear proceeds by atomic layer removal.—K. E. J.

Metal Transfer During Static Loading and Impacting. E. Rabinowicz. (*Proc. Phys. Soc.*, 1952, **65B**, Aug. 1, 630–640). Experiments were made on transfer of metal (Cu, Cd, Ag, and stainless steel) when radioactive hemispherical sliders are pressed into flat surfaces. Metal transfer as small fragments is always present, supporting the view that metallic junctions are formed. At higher loading, the surface oxide film is broken up to a greater extent, giving more than a proportional increase in metallic interaction and transfer : an analogous result is produced with boundary lubricants. Similar results were found from impacting, though pick-up was reduced. With lubricants a trapped surface may further reduce pick-up without affecting plastic deformation. (10 references).

Westinghouse Electric Correlates Its Product Development with Research. J. Parina, jun. (*Metal Progress*, 1952, **62**, Aug., 73–78). General progress in research and development is reviewed and the close co-ordination between these two departments is stressed. Examples are given of several alloys which have been discovered, including Hipernik (Fe 50%, Ni 50%) which is being used in magnetic amplifiers. Work on heat-resisting alloys led to an improvement in investment casting, techniques and the new ' hot cold-working ' technique. Among other investigations in progress are studies on the effect of gases in metals and notched-bar strengths at high temperatures. Future lines of research are proposed.—B. G. B.

An Automatic Plotting Machine for Magnetization Curves. E. Rawlinson. (*Metro Vick Gaz.*, 1952, **24**, Oct., 257–262).

Apparatus for Measuring the Total Energy Losses in Magnetic Sheet. M. P. Andres Sanz. (*Inst. Hierro Acero*, 1952, **5**, Aug., 864–873). [In Spanish]. An apparatus is described for measuring the total energy loss due to hysteresis and Foucault currents in magnetic sheets subjected to alternating magnetic fields (50 cycles/sec.), and with maximum magnetic inductions up to 18,000 gauss, using small samples of up to 2 kg.—R. S.

Magnetostriction of the Ferrites of Cobalt as a Function of the Composition. R. Vautier. (*Comptes Rendus*, 1952, **235**, Aug. 4, 356–358).

The Magneto-Resistance of Ferromagnetic Al–Si–Fe Alloys. R. Parker. (*Proc. Phys. Soc.*, 1952, **65B**, Aug. 1, 616–620). By extending the treatment in an earlier paper, an equation is derived for the saturation magneto-resistance of some mixed ferromagnetic alloys as a function of composition and temperature. Results calculated from this equation are in good agreement with experiment for the Al–Si–Fe alloy system.

Some Post-War Developments in Magnetism. L. F. Bates. (*Proc. Phys. Soc.*, A, 1952, **65**, Aug. 1, 577–594). A survey is given of some of the more outstanding developments in magnetism during the post-war years, including the extension of knowledge of ferromagnetic domains through the theoretical work of Néel and others, and from the systematic use of the Bitter powder figure techniques. The main ferro-magnetic, para-magnetic, and nuclear resonance phenomena, and experiments on the diffraction of neutrons by anti-ferromagnetic crystals are also described. (18 references).—K. E. J.

Temperature Dependence of Magnetic Relaxation of Iron–Nickel Alloys. E. F. Kuritsina. (*Doklady Akademii Nauk S.S.S.R.*, 1952, **84**, 4, 687–688). [In Russian].

Structure Transition and Antiferromagnetism in Magnetite. N. C. Tombs and H. P. Rooksby. (*Acta Crystallographica*, 1951, **4**, Sept., 474–475). [In English]. Results of an investigation on the crystal structure of magnetite at temperatures down to 95° K. are reported. The analysis of the X-ray powder photographs obtained is described, and it is concluded that the low-temperature structure transition is associated with an antiparallel orientation of the spins of magnetic atoms.

Physical Structure and Magnetic Anisotropy of Alnico 5. R. D. Heidenreich and E. A. Nesbitt. (*J. Appl. Phys.*, 1952, **23**, Mar., 352–365, 366–371). In the first part, electron-diffraction and microscope evidence is given showing that the high coercive force and anisotropy of Alnico 5 are due to a very finely divided precipitate produced by the permanent magnet heat-treatment. The precipitate includes cobalt, and its shape is described. Domain rotations rather than boundary movements account for the coercive force of this alloy ; this is discussed in the second part with reference to various theories.—E. T. L.

Memorandum on Non-Destructure Methods for the Examination of Welds. (*Brit. Welding Res. Assoc.*, 1952, T.29.)

Details of several methods of examination are given. The scope and use of individual methods and the apparatus required are discussed in detail. The methods discussed include radiography, magnetic, fluid penetration, electrical (including induction, resistance, and eddy-current techniques), chemical, and acoustical (including sonic and ultrasonic methods). The applicability of these methods for the examination of 25 different types of welded joint is given together with a description of the type of weld defect revealed by the examinations.—B. G. B.

Ultrasonics—Inspection Applications Grow. R. H. Frank. (*Steel*, 1952, **130,** June 16, 99–102). The principles involved in the ultrasonic system of testing for internal flows in castings and forgings are explained. The value of the tests depends upon the instrument operator and upon careful interpretation of the data obtained. A number of examples of different techniques and the results to be expected are given. Trends now are towards automatic handling equipment for repetitive testing.—A. M. F.

Curved Crystal Developments in Ultrasonic Resonance Testing. C. R. Betz. (*Non-Destructive Test.*, 1952, **10,** Spring, 28–31). In order to obtain better contact, and thereby better test results on round bars and curved surfaces, the author has experimented with quartz crystal probes ground to fit approximately the surface in question. The results of such tests show improvements over flat crystals, and the curved varieties are now obtainable from the Magnaflux Corp. Typical applications of the new probes are described.—P. M. C.

Ultrasonic Inspection Checks Quality of Brazed Joints. H. Greenberg. (*Mat. Methods*, 1952, **35,** June, 102–105). Good correlation has been obtained between the results of shear testing and ultrasonic inspection of brazed circuit-breaker contacts. The ultrasonic test has accordingly been adopted as the acceptance test for these parts. The procedure and examples of the correlation with shear values are given.

Magnetic Properties, Internal Strains, and the Bauschinger Effect in Metals. D. V. Wilson. (*Nature*, 1952, **170,** July 5, 30–31). The mechanism of work-softening effects is discussed in relation to the nature of the textural stress systems present. Results for a heat-treated, 0·88% carbon steel with a maximum softening effect corresponding to an initial hardness of about 600 Vickers hardness are given together with a possible explanation.—A. G.

Ultrasonic Measurement. (*Elect. Times*, 1952, **122,** July 17, 101–104). The theory of ultrasonic measurement of wall thickness by the resonance method is described. The Dawe ultrasonic thickness gauge has been designed on this principle and operates in the range of $\frac{1}{16}$ in. to 12 in. An accuracy of about 1% is obtainable on most surfaces, except those which are heavily scaled.—H. D. W.

Advances in Application of Ultrasonic Vibrations. (*Times Rev. Ind.*, 1952, **6,** July, 30–33). The applications of ultrasonics in many fields are mentioned, including measurements of the elastic properties of iron from −80° C. to as high as 1000° C., breaking down the oxide film on aluminium to allow soldering, and the coagulation of dusts in gas cleaning. The use of ultrasonic waves in an echo system for detecting flaws in castings and forgings is also surveyed. Other applications include the location of objects in the dark and the quartz crystal clock.—T. E. D.

X-Ray Perfect. D. Goodman. (*Machinist*, 1952, **96,** May 31, 827–829). Four commercial X-ray laboratories in Chicago are mentioned and the work of two of them is briefly described.

Radiographic Inspection of Materials. (*Instrument Practice*, 1952, **6,** Apr., 380–387). The methods of radiography applied to the examination of welding in ships is described, and examples are given of the use of the Solus-Schall mobile X-ray equipment. The development of gamma rays instead of X-rays is discussed, and examples are given of the relative merits of the two sources.—H. D. W.

Multi-Directional Magnetic Particle Inspection. R. A. Peterson. (*Non-Destructive Test.*, 1952, **10,** Spring, 18–22). Until recently, the detection of transverse and longitudinal defects by magnetic particle methods has necessitated two operations, one circularly and one longitudinally magnetized. The Duovec multidirectional method of magnetization, developed by the Magnaflux Corp., now enables both types of defect to be revealed in one operation. The method is described, and equipment and typical results are shown.—P. M. C.

Study of Magnetic Fields at a Crack in a Magnetized Steel Plate. R. D. Kodis and G. A. Darcy. (*A.S.T.M. Bulletin*, 1952, Apr., 71–73). A crack of varying depth was purposely produced in a steel plate, and the magnetic field leakage around this crack was studied by means of a magnetic tape-recording head. The magnetic leakage field was measured with respect to the crack depth level of magnetizing force applied to the steel plate, and distance of the search head above the cracked plate. Continuous magnetization was more sensitive than residual magnetization for the detection of cracks.—B. G. B.

White Magnetic Fluid for Crack Detection. J. W. Walley. (*Metro. Vick. Gaz.*, 1952, **24,** June–July, 201). The development of a white magnetic fluid using finely divided aluminium as the additive is described. Aluminium powder does not affect the magnetic sensitivity of the fluid, and suspension of the particles can be sustained.—B. G. B.

Selection of X-Ray Detectors for Automatic Inspection. J. E. Jacobs and A. L. Pace. (*Non-Destructive Test.*, 1952, **10,** Spring, 12–17). The detection and recording of X-radiation depends usually upon ionization of gases, phosphorescence of phosphorus, or resistivity changes in semi-conductors. Ionization chambers, Geiger-Müller tubes, and the use of photomultiplier tubes are discussed. Finally, a fourth type of detector, a photoconductor which shows high dark resistance and very high amplification of the primary photocurrents, is hexagonal cadmium sulphide. The use of such crystals for X-ray detection is described in detail. (14 references).

A Procedure for Reproducing Radiographs. D. T. O'Connor and L. V. Burt. (*Non-Destructive Test.*, 1952, **10,** Spring, 9–11). A procedure is described which enables faithful copies of radiographs for final paper prints to be made. The procedure depends upon the contrast variations obtainable with commonly available films and developers. A whole series of test negatives is prepared using different films, developers, and times, and the desired contrast range for a specific X-ray reproduction is measured by a densitometer and matched with one of the test series. This series is, of course, standard for a particular laboratory and equipment.—P. M. C.

Various Inspection Methods Used in Heat Treating Shops. S. A. Gill. (*Metal Treating*, 1952, **3,** May–June, 9–11). Magnaflux magnetic particle crack detection methods are first described, and the newly developed Zyglo method suitable for non-ferrous metals is also dealt with.—P. M. C.

Miniature Radiography. S. L. Fry. (*Metal Ind.*, 1952, **80,** June 13, 479, 480). In order to cut down the high film cost involved in contact radiography, a technique has been developed whereby a normal fluorescent screen, as used for visual examination, is photographed by a miniature camera of 35-mm. film. Examination of the negatives by means of a film-strip projector is just as good as visual screening, with the added advantage that a permanent record is available. The equipment is described.—P. M. C.

Industrial Radiography. C. W. Jackson. (*J. S. African Inst. Mech. Eng.*, 1952, **1,** Apr., 264–282). In a survey of industrial X-ray inspection techniques, the practical aspects of radiation scatter and high and low voltage operation are considered, and films, film holders, and lead and fluorescent screens are discussed. Radioactive isotopes as radiation sources are compared with X-ray generators, and it is concluded that the latter are superior for low-voltage inspection up to about 3 in. of steel. For heavy sections, each method has its advantages and radioactive isotopes are preferred in view of lower cost if long exposure times can be tolerated.

Radiography as an Aid to Research. A. W. Balls. (*G.E.C. J.*, 1952, **19,** July, 177–185). The penetrations of various metals and carbon by X-rays (100–1000 kV.) is given. Radiography techniques are described and the use of a gamma-ray source is mentioned ; some radiographs are shown.—T. E. D.

The ERESCO 175-kV, 5-mA Industrial X-Ray Unit—A Portable Apparatus for Testing Materials. E. Mittli. (*Betrieb. u. Fertigung*, 1952, **6,** June, 108–110). This X-ray equipment and the technique for using it are described, and a chart for determining the correct exposure is presented.

Radiography in Engineering. L. Mullins. (*J. Inst. Prod. Eng.*, 1952, **31,** June, 265–284). A number of applications of radiographic methods to engineering problems are described. The use of X-ray and gamma-ray sources are discussed. Examples of the examination of welds and castings by radiographic methods are given.—B. G. B.

Natural and Artificial Sources for Gamma Radiography. B. N. Clack. (Inst. Physics : *Engineer*, 1952, **194,** Aug. 8, 195–198). The author considers the requirements of a source

suitable for industrial gamma radiography and lists some 20 elements giving the half-lives and emission energies. The author also describes the preparation of a radiographic source using radon, and explains some of the problems met with in doing this.—M. D. J. B.

Gamma-Radiography Control of Steel Castings. R. Duchenne. (*Arts et Manufactures*, 1952, July, 13–16). Cobalt^{60} is particularly suitable as a gamma-ray source because of its convenient half-life period, the penetrating power of its radiation, and its low cost. Examples of practical tests are given, with details concerning technique.—T. E. D.

A Transient Heat Flow Method of Determining Thermal Conductivity: Application to Insulating Materials. C. P. Lentz. (*Canad. J. Techn.*, 1952, **30,** June, 153–166). A line heat source as a probe was found unsatisfactory for use in insulating materials. An unenclosed wire heater gave good results with granulated and blanket rock wool, silica aerogel, and slab cork. Conductivities obtained varied linearly with heater current; they were not significantly affected by the position of the heater wire, horizontal or vertical.

Method of Determining the Coefficients of Thermal Diffusivity and Conductivity of Steels. N. Iu. Taits and E. M. Gol'dfarb. (*Zavodskaya Laboratoriya*, 1950, No. 3, 314–319). [In Russian]. A method directly applicable to the determination of the thermal diffusivity and conductivity of large castings is described; it gives results sufficiently accurate for the calculation of heat-treatment conditions. It consists of heating flat, cylindrical, or square prism specimens in any furnace giving symmetrical heating, and measuring the temperatures at the surface and centre of the specimen. The temperature differences at the start and end of each heating period are treated graphically to give the required coefficients.

Physical Principles of Creep. P. Laurent. (*Métaux-Corrosion-Indust.*, 1952, **27,** June, 265–273). A review is made of creep studies. The deformation produced during creep is examined and the effect of several variables (including the applied stress, temperature, chemical composition) on the rate of creep is discussed. The actual rupture produced by creep and the metallurgical condition of the specimen on rupture are reviewed.—B. G. B.

Problems Posed by the Use of Refractory Materials in Gas Turbines. R. Fleury. (*Métaux-Corrosion-Indust.*, 1952, **27,** Apr., 160–163). The severe conditions under which turbine parts, the combustion chamber, and distributors operate in a gas turbine are reviewed and the use of ceramics for their manufacture is discussed. The difficulties involved in the manufacture and use of ceramic parts are reviewed.—B. G. B.

General Trends in the Development of Materials for Gas Turbines in Great Britain. A. H. Waterfield. (*Métaux-Corrosion-Indust.*, 1952, **27,** Apr., 156–157). A short account is given of recent work on high-temperature alloys and sintered powders for use in gas turbines.—B. G. B.

Ceramics Shrug off Turbine's Tempest. J. V. Long. (*S.A.E. J.*, 1952, **60,** Sept., 38–41). A new refractory ceramic coating for use in jet engines has been developed by Solar Aircraft Co., U.S.A. Details of this 'Solaramic Process' are not divulged, but test results on coated type 321 stainless steel are reported. These show that this coated steel withstands temperatures of 1700 to 1900° F. just as well as Inconel.—P. M. C.

Stainless-Clad Copper Promises Improved High-Temperature Service. (*Machine Design*, 1952, **24,** June, 141–142). **Clad Copper Sheet.** (*Metal Ind.*, 1952, **81,** Aug. 15, 125, 126). Brief details are given of 'Rosslyn Metal' developed by the American Cladmetals Company. The laminated sheet, of copper faced with thin stainless outer layers, has a comparative thermal conductivity of 57·2% (based on copper at 100%) as against 6·1% for stainless steel alone. The material is being tried in many fields including jet-engine components and exhaust manifolds.—P. M. C.

Stainless Clad Copper for High Temperature Service. J. Kinney. (*Product Eng.*, 1952, **23,** Apr., 129–133). This material offers for high temperature service a combination of the high thermal conductivity of copper with the corrosion and heat resistance of stainless steel. The highest lateral conductivity is attained with about a 50% copper core.

New High-Temperature Alloys. A. R. Edwards. (Australian Inst. Metals: *Australasian Eng.*, 1952, June 7, 42–47). The author discusses the reasons for basing future high-temperature alloys on the Group VIA metals of the Periodic Table, *i.e.*, chromium, molybdenum, and tungsten. He discusses cohesion in the transition metals, and outlines the major developments in, and mechanical properties of, alloys containing the above-named metals as the main constituents. (29 references).—P. M. C.

Correlation of Creep Properties by a Diffusion Analogy. L. Green, Jun. (*J. Appl. Mech.*, 1952, **19,** Sept., 320–326). By considering creep as essentially self-diffusion, the author proposes an empirical correlation between data for constant-stress creep and for diffusion, in an attempt to devise a method of deducing the former from the latter. Data so far available show a similar temperature dependence for the activation energies of both processes, but the correlation obtained is not sufficiently accurate to be useful, and the analogy will not hold when slip is the creep mechanism.—E. T. L.

Some Properties of a Nodular Iron at Elevated Temperatures. M. S. Saunders and M. J. Sinnott. (*Amer. Soc. Met. Preprint* No. 19, 1952). Data are presented giving the tensile strength between 425 and 650° C. of nodular iron, either as-cast or partially annealed. Except for ductility and high-temperature impact strength, the nodular iron is better than carbon–molybdenum or killed carbon steels, in respect of short-time tensile properties.—E. T. L.

Relation Between the Results of Creep Tests and Fatigue Tests at High Temperature on Certain Steels. B. Andreini and A. Erra. (*Met. Ital.*, 1952, **44,** Aug.–Sept., 299–307). The authors establish that at high temperatures there is interaction between the effects of creep and fatigue. This interaction is a function of the cycle frequency and the duration of the test. The fatigue tests and the testing equipment are described in detail. Experimental results, and microphotographs showing the importance of surface corrosion by oxidation, are given. Corrosion starts at the grain boundaries and forms a starting point for fatigue fractures. The creep tests were continued until fracture took place. Failure–time curves are shown. At the test frequencies examined, it is always the alternating load which is responsible for the ultimate failure of the material even in the presence of creep.—M. D. J. B.

Creep Properties of Steels Utilized in High-Pressure and High-Temperature Superheater and Steam Pipe Practice, Part IV—Revision of Long-Time Creep Data for a Carbon Steel Header and a Carbon–Molybdenum Steel Pipe. R. W. Ridley. (*Inst. Mech. Eng. Advance Copy*, Dec. 2, 1952). The author gives revised temperature estimates for a 0·24% C superheater header steel at 5 tons/sq. in., and for a 0·09% C, 0·50% Mo steel at 6 tons/sq. in., based on the results of creep tests at these stresses lasting 67,000 hr. The extended tests indicate that modifications are necessary in the previously extrapolated temperatures corresponding to 100,000 hr. The tests are being continued for the full period of 100,000 hr.

Effect of Cold Drawing on the Creep Resistance and Structure of 0·5 per cent. Molybdenum-Steel Steam Piping. E. A. Jenkinson and D. C. Herbert. (*Iron Coal Trades Rev.*, 1952, **165,** Aug. 15, 363–368: *Brit. Elect. Allied Indust. Res. Assoc. Technical Report* J/T154). See *J. Iron Steel Inst.*, 1952, **171,** July, 334.

Creep and Rupture of Chromium–Nickel Austenitic Stainless Steels. E. J. Dullis, G. V. Smith, and E. G. Houston. (*Amer. Soc. Met. Preprint No.* 7, 1952). The creep properties of several 18/8 stainless steels, with and without additions of molybdenum, niobium, or titanium, after different forms of heat-treatment were investigated. Some grades with only 0·03% max. carbon are significantly weaker in creep than the same grades higher in carbon. Sigma formation and carbide precipitation were the principal microstructural changes observed.—R. A. R.

Iron and Steel. A. M. Sage, J. Pearson, and J. C. Hudson. (*Soc. Chem. Ind., Reports on the Progress of Applied Chemistry*, 1951, **36,** 233–263). Recent advances in the development of steels for special purposes are reviewed with particular reference to high-temperature alloys for gas turbines, magnet alloys, and tool steels. Boron steels for constructional purposes are mentioned. Heat-treatment and surface treatment are surveyed. The advances in fundamental process chemistry are outlined, with special reference to desulphurization and deoxidation in the iron- and steelmaking processes. The great importance of fighting corrosion, and recent trends in the study of the problem are indicated. Work on atmospheric, wet, and soil corrosion is referred to, and cathodic protection and various coatings are mentioned. (230 references).

Metal–Ceramic Composites. J. H. Westbrook. (*Bull. Amer. Ceram. Soc.*, 1952, **31,** June, 205–208 July, 248–250).

This paper reviews the general field of metal–ceramic materials with respect to their history, applications, forms, properties, manufacture, and design.—E. C. S.

The Physics of the Deformation of Metals. E. N. da C. Andrade. (*Rév. Mét.*, 1952, **49**, July, 469–484). An apparatus maintaining constant stress during creep tests reveals the existence of two distinct forms of creep, transient and permanent. The effects of load and temperature on these and of recrystallization in producing accelerated creep are discussed. Experiments in which a specially shaped thin annulus is tested in simple shear are described. The use of monocrystals to simplify study is considered and their main properties outlined. Finally, the influence of surface is discussed.—A. G.

The Problem of Heat-Resisting Alloys in the Construction of Automobiles. H. Godfroid. (*Métaux-Corrosion-Indust.*, 1952, **27**, Apr., 158–159). A number of engine parts operate at temperatures above 400° C. The use of alloy steels for the construction of valves, sparking-plug electrodes, and the reduction of the wear of moving parts is discussed.—B. G. B.

Conditions of Appearance of Rupture During Study of Creep. G. Vidal. (*Métaux-Corrosion-Indust.*, 1952, **27**, May, 201–215). Factors causing acceleration of creep, including change in the dimension of the sample, presence of microfissures and crystalline rearrangements, are reviewed. The actual initiation of the rupture and the physical and metallurgical changes which occur at this point are next discussed. The influence of several variables (including temperature, speed of deformation and oxidation) and the effect of initial stresses, grain size, temperature, oxidation, and sample dimensions on the hardness before rupture, are described in detail.—B. G. B.

Equipment for Creep Testing at Temperatures up to 200° C. R. J. M. Payne. (*Metallurgia*, 1952, **45**, June, 322–324). A detailed description is given of equipment that enables six tests to be carried out simultaneously at the same temperature. The time taken in adjusting the temperature and the temperature uniformity of creep-test specimens is greatly reduced by heating them in an oil bath. Strain measurements were made using micrometers, but optical extensometers could be used with the equipment. Shell Voluta oil was found suitable for running at 200° C. but below 100° C. a less viscous oil is desirable.—B. G. B.

Contribution to the Study of Pipes Submitted to High Temperatures. H. Carlier. (*Chaleur et Ind.*, 1952, **33**, May, 139–150). The expansion and forces in pipes submitted to high temperatures are considered and treated mathematically. Strains in various shaped pipeline systems are evaluated.

Steam Pipe Materials for Advanced Steam Conditions. D. W. Crancher. (*Trans. Inst. Marine Eng.*, 1952, **64**, 147–165). Materials suitable for steam pipes working up to 1050° F. are surveyed. Details are given of properties of 0·15% carbon steel, and three low-alloy steels containing 0·5% Mo, 0·54% Mo with 0·265% V, and 0·7% Cr with 0·5% Mo respectively. Stress–temperature curves, giving 0·5% creep after 100,000 hr., and also to rupture after 100,000 hr., are shown. These data are discussed fully and recommendations are given for the appropriate working range for each steel ; the limit for low carbon (normalized) steel is 850° F., while the most resistant is Mo–V steel (normalized and tempered), which is suitable for the range 950–1050° F. An appendix includes notes on creep of metals and creep testing. (55 references).—T. E. D.

On the Problem of Studying the Properties of Metals and Alloys at High Temperatures *In Vacuo*. N. T. Gudtsov, M. G. Lozinskii, I. F. Zudin, N. A. Bogdanov, and M. P. Matveeva. (*Izvestiya Akademii Nauk S.S.S.R., Otd. Tekh. Nauk*, 1950, No. 1, pp. 108–124). A description of an apparatus for the micrographic analysis of metals and alloys during heating *in vacuo* is given. It may also be used for studying hardness at temperatures up to 1000° C., resistance to corrosion, the kinetics of formation of oxide films, and the process of deformation in heated samples.—V. G.

Conservation of and/or Substitution for Critical Jet Engine Materials. N. E. Promisel. (*J. Met.*, 1952, **4**, July, 698–702). To meet the specific requirements of different sections and components of a jet engine, many different materials and alloys are used. In view of the acute shortage of some of the critical materials, the author reviews possible means of conservation or substitution.—G. F.

Steel Without Molybdenum for Steam Turbines. S. Todl. (*Strojírenství*, 1952, **2**, No. 6, 253–260). A critical survey of various methods of creep testing and of the interpretation of creep data is given. A metallurgical evaluation of quality, on the basis of the creep properties of alloy steels used for turbines and boilers, is attempted. Creep data and permitted creep stresses as given in Czech and American standard specifications, are shown. The stress distribution in a turbine rotor is calculated as a basis for the choice of suitable steels. It is shown that designers have not yet a correct conception of the properties of steels above 500° C. The only really effective elements for increasing creep strength are molybdenum and titanium ; the series of elements (Mo, W, Ti, V, Nb, Cr, Zr, Mn, and Ni) that Bennek and Bandel showed to be of decreasing effectiveness (in the order given), remains practically in the same order even above 500° C. Cr–Ti and Si–Ti steels have higher creep strengths than molybdenum steels, and thus titanium can be used to replace molybdenum.—P. F.

Some Turbine Problems in the Development of the Whittle Engine. Sir Frank Whittle. (Fifth Hatfield Memorial Lecture : *Symposium on High-Temperature Steels and Alloys for Gas Turbines, Iron Steel Inst. Special Report* 43, 1952, 379–386). The author briefly sketches the story of the early development of the Whittle engine, with emphasis on the turbine problems encountered.—R. A. R.

Thermal Shock Tests on Gas Turbine Materials. M. Bentele and C. S. Lowthian. (*Aircraft Eng.*, 1952, **24**, Feb., 32–38). A general theoretical investigation is made of thermal stresses in materials brought about by static and transient conditions. Published results of thermal shock tests are outlined, and two new test methods are described for nozzle blades in which cyclic, or cyclic and steady conditions can be applied.—T. E. D.

High Temperature Alloys. J. Convey. (Symposium on Metallurgical Research in Canada, 1951, April ; *Canad. Min. Met. Bull.*, 1952, **45**, Apr., 217). Research in Canada shows it is likely that titanium-base alloys can be developed for service up to, but not above, 1000° F. A new precision casting high-temperature alloy, Kinsalloy, has been developed with the nominal composition 70% Ni, 22% Mo, and 8% Al. Work on improving the oxidation resistance of this alloy and modifying it to contain less molybdenum is continuing.

Creep Resistance of Carbon–Molybdenum Steels. W. Tomaszczyk and Z. Borysowski. (*Prace Instytutu Metalurgii*, 1952, **4**, 2, 161–176). [In Polish]. Factors affecting the creep strength of carbon–molybdenum steels, such as method of production, chemical composition, and heat-treatment are surveyed. On the basis of results in the literature, mean values of the mechanical properties of 0·5% Mo steel at high temperatures are given. (64 references).—V. G.

Improvements in Heat-Resistant Chromium–Molybdenum Steels. O. L. Bihet and A. Gourlez de la Motte. (*Ossature Métallique*, 1952, **17**, June, 327–331). Creep and corrosion resistance of a 2·25% Cr 1% Mo steel and their dependence upon heat-treatment and small additions of vanadium were investigated., Optical and electron micrographs were taken of the structural changes accompanying heat-treatments, and measurements of the creep limits and tensile strength were made. Normalizing at 930° C., followed by a 6-hr. recovery treatment at 775° C. (25° C. below the Ac_1 point) gave best results. Increased hardness was obtained by replacing about one fifth of the molybdenum by vanadium.—P. F.

The Behaviour of Metals at Low Temperatures. F. Pambieri. (*Calore*, 1952, **23**, Aug., 392–396). The author examines the principal mechanical characteristics of metals at temperatures in the region of −200° C. The tendency of most metals, at low temperatures, to lose malleability and plasticity and to become more brittle is confirmed. Copper and copper alloys are little affected by cold temperatures. Ferrous metals are affected, particularly when they have been submitted to certain types of heat-treatment. The behaviour of welds in ferrous metals may be uncertain at these low temperatures.—M. D. J. B.

Selecting Metals for Low Temperature Service. J. R. Watt. (*Product Eng.*, 1952, **23**, June, 158–161). The selection of steel for low-temperature application requires particular consideration ; factors affecting the transition temperature being grain size, carbon content, heat-treatment, and welding temperatures. The improvements of nickel additions on impact resistance for various notch sharpnesses are plotted, and they are recommended for ultra-low-temperature use. Other critical factors are impact velocity, section thickness, presence of notches or stress raisers, and residual stresses.—A. M. F.

Materials of Construction for Use at Low Temperatures. E. W. Colbeck. (*Trans. Inst. Chem. Eng.*, 1949, **27**, 55–69).

Low-temperature embrittlement is explained, and the behaviour of ferritic steel at low temperatures is outlined. The effects of microstructure, deoxidation practice, and alloying elements are surveyed. Austenitic steels are also examined, and microstructures are illustrated. The behaviour of some non-ferrous metals is also mentioned.—T. E. D.

Economic Use of Alloy Steels. L. Jeníček. (*Strojírenství*, 1952, **2**, 6, 250–253). [In Czech]. The most rational and economic use of alloy steels in constructional and machine building applications is discussed. The simplification and standardization of alloy steel production is considered, and possible changes in the steel nomenclature current in Czechoslovakia are examined.—P. F.

Rare Earths Improve Properties of Many Ferrous Metals. W. E. Knapp, and W. T. Bolkcom. (*Iron Age*, 1952, **169**, Apr. 24, 129–134 ; May 1, 140–143). This article presents information on the use and effects of ' Lan-cer-amp,' a trade mixture of rare earths, and reviews the history of somewhat comparable additions. Ladle additions of ' Lan-cer-amp,' which contains rather more lanthanum than regular misch metal, are made to stainless, tool, alloy, and electrical steel grades at the rate of $\frac{3}{4}$ to 6 lb./ton. Although the rare earths are strong desulphurizers not all the benefits derived from their use can be attributed to sulphur reduction. The sulphur drop experienced after the addition is only indicative that the metal has been properly treated. The effect of the addition on nitrogen content may be even more important. Treated steel castings show higher impact and ductility in quenched and tempered condition than do untreated heats. Transverse impact values at −40° F. on 0·32% carbon Cr–Ni–Mo forgings were increased from 12 ft.lb. to 20 ft.lb. through rare-earth additions. Increased fluidity of treated metal is characteristic, and a 35° F. drop in the solid–liquids temperature has been noted. Less grain coarsening in austenitic stainless steels, type 310, is experienced when 2 lb. of Lan-cer-amp per ton is used. (28 references).—A. M. F.

Alloy Steel Conservation. (*Nickel Bull.*, 1952, **25**, May, 125–127). **Nickel and Molybdenum in Alloy Steel.** (*Metal Treatment and Drop Forg.*, 1952, **19**, May, 203). **Economy in Steel Alloying Elements.** (*Metallurgia*, 1952, **46**, July, 31–32). The Ministry of Supply's direction restricting the use of nickel and molybdenum is outlined. Tables are given that show the maximum percentages of each or both of the elements which may be used for products having the minimum tensile strength and ruling section specified. Further tables list En and other steels complying with the Ministry's direction.

An Investigation of Boron and Other Low-Alloy Steels. H. B. Knowlton. (*Steel Processing*, 1952, **38**, Feb., 72–75 ; Mar., 131–133 ; Apr., 186–190). The need for scarce alloy conservation is discussed, and the possibility of using boron and ' leaner ' alloy steels is dealt with. Compositions of such steels, and the suggested substitutions for standard grades are tabulated. The requirements for alternative steels and the more usual causes for breakage are discussed. The Jominy test and hardenability curves for selecting suitable steels for specific jobs are discussed, and the use of certain boron steels for case-hardened parts is dealt with.—P. M. C.

Distortion and Service Tests of Carburized Gears Made of Boron Steels. S. L. Widrig and W. T. Groves. (*Metal Progress*, 1952, **62**, July, 75–78). Owing to excessive bore shrinkage and tooth distortion, boron steels are unsuitable for the manufacture of ring gears. The martempering of gears of boron steel has been successfully carried out. The results of fatigue and chip resistance tests carried out on a number of American boron steels are given.—B. G. B.

Chrome Stainless: Easing The Pinch. R. E. Paret. (*Steel*, 1952, **130**, Apr. 28, 79–80). Stainless steels with 12%, 17%, and 27% Cr can be used for applications formerly requiring nickel–chromium steels. The physical and mechanical properties and the effects of temperature on the properties of these types are given.—A. M. F.

Forgeability, Machinability and Hardenability of Boron Carburising Steels. S. L. Widrig and W. T. Groves. (*Metal Progress*, 1952, **61**, June, 71–76). Boron has no detrimental effect on forging properties. Machinabilities were compared through operations such as turning, broaching, and milling ; no significant differences were observed. Hardenability of core and case was examined ; the chief limitation of boron was in the decrease in effectiveness as the carbon content of the steel increases.—B. G. B.

How to Substitute a Boron Steel. H. B. Knowlton. (*S.A.E. J.*, 1952, **60**, May, 74–77). The author discusses hardenability as the criterion for the suitability of steel for specific applications. The Jominy test and its meaning are dealt with, and the uses of boron steels for automobile parts are given in a table.—P. M. C.

Copperweld Fine Wire. F. E. Leib. (*Wire and Wire Products*, 1952, **27**, Sept., 878–880, 927). Mechanical and electrical properties of wire made from copper-clad steel are quoted, and applications in the electrical, radio, and musical industries are instanced.—J. G. W.

Cobalt: Its Industrial Applications. J. B. Richardson. (*Times Rev. Ind.*, 1952, **6**, Sept., 22–27). The wide uses of cobalt and its alloys are surveyed in some detail. The compositions of high-speed cutting steels containing cobalt are included. Magnets with 0·5–35% Co are mentioned. The application of stellites where wear resistance is essential is outlined, and carbides for tool bits are dealt with. The use of Cobalt^{60} for gamma radiography is described, and variou minor uses are reviewed.—T. E. D.

Selection of Steels for Gear Wheels: Derivable Physical Properties in the Metal. C. Thielers and S. G. Klemming. (International Mechanical Engineering Congress, Stockholm : *Iron Coal Trades Rev.*, 1952, **165**, Aug. 1, 267–269). The authors consider the selection of carbon and low-alloy steels for the production of traction motor gears, and they discuss Swedish views on heat-treatment methods and on the optimum hardness distribution in the gear teeth.—G. F.

The Fabrication of Chemical Plant in Stainless Steel. A. C. Branch and J. L. Sweeten. (*Trans. Inst. Chem. Eng.*, 1949, **27**, 139–167). The fabrication of stainless steel is reviewed. Cutting, machining, deforming, welding, heat-treatment, descaling, finishing, and other processes are described. The availability of alloys, and its influence on design in Great Britain are discussed. (35 references).—T. E. D.

Russian Tractors Show Sound Engineering. W. G. Patton. (*Iron Age*, 1952, **169**, June 5, 150–152). This report gives some of the findings of an engineering and metallurgical examination of two captured Russian tractors. Nickel and chromium have been substituted for molybdenum, and induction hardening has been applied. Close machining tolerances are held where necessary, otherwise very high speeds and feeds are used in machining. Considerable use of castings is made, but internal cleaning of castings is poor.—A. M. F.

Industrial Blades and Knives: Manufacture in France. (*Iron Steel*, 1952, **25**, Aug., 358). The author discusses in general terms the production of mechanical cutting machines, which are used in a wide range of industries. It is stressed that the characteristics of the material to be cut, the output rate, and the condition of the machines to which the blades are fitted, must all be taken into account when selecting the blade material.—G. F.

Transition from Ductile to Brittle Behaviour in Pressure Vessel Steels. T. N. Armstrong, N. A. Kahn, and H. Thielsch. (*Welding J.*, 1952, **31**, Aug., 371s–380s). This is an extensive review of data on metallurgical properties of welded pressure-vessel steels. (97 references.)—U. E.

Comparison of En Steels with other Official Specifications. (*Nickel Bull.*, 1952, **25**, Mar., 54–57). A comparison of En steels with other official specifications is provided in tabular form. The data cover the United Kingdom, U.S.A., and most European countries. The tables also include steels, which, though not similar in chemical composition, are possible alternatives.—P. M. C.

Use of Needle Tubing Extended to Mechanical Applications. (*Mat. Methods*, 1952, **35**, June, 106–108). The high strength, stiffness, and strict dimensional tolerances that characterize this tubing open new fields for a product originally designed for surgical uses. The properties of needle tubing are discussed and tabulated.—P. M. C.

How Good is 80B30? R. N. Imhoff and J. W. Poynter. (*Iron Age*, 1952, **169**, June 26, 102–107). The temper brittleness, fatigue, and low-temperature impact strength of boron-treated low-alloy steels in the range of 0·30 to 0·45% C have been studied by the U.S. Air Force. Where Charpy impacts of 10 to 20 ft.lb. are considered adequate, 80B30 can be used in the normalized condition down to −100° F. At high strength levels, it shows good impact strength but was found be subject to temper brittleness.—A. M. F.

Free-Machining Steels. C. Vouga. (*Rev. Mét.*, 1952, **49**, June, 393–407). A brief account of the development of free-

machining steels is followed by a discussion of the effect of structure and the metallurgy of such steels. Their classification is illustrated by tables, and factors governing the uses and choice of steel type are considered.—A. G.

Guide for Selecting Tool Steels and Carbides. (*Steel*, 1952, **130,** June 16, s-1–s-32). This is a special report by the editors of *Steel* giving a list of more than 50 companies and the tool steels available from them. Details are given of the primary applications of the steels, their trade names, the chemical composition, quenching medium, machinability annealed, and the movement in hardening.—A. M. F.

Tool Steels: Uses, Types, Treatments. T. C. DuMond. (*Mat. Methods*, 1952, **36,** Aug. 113–128). This is a descriptive article covering classes of tool steels, factors to be considered in their selection, and heating and heat-treatment.—P. M. C.

Factors in Selecting Stainless Steels. B. T. Lanphier. (*Machine Design*, 1952, **24,** July, 112–128). Illustrated case histories are presented to show how a particular stainless type may be adapted to a specific set of requirements so as to reduce costs, improve performance, and conserve scarce alloying elements needed for the defence programme.—E. C. S.

Cold Finished Carbon Steel Bars. (*Amer. Iron Steel Inst., Steel Products Manual Section* 9, Sept,. 1952). This 78-page booklet makes available to purchasers, producers, and consumers information on manufacturing practices, quality descriptions, sizes produced, and properties of cold-finished steel bars manufactured in the U.S.A.—R. A. R.

High Chromium Stainless as a Substitute for 18/8. L. F. Spencer. (*Steel Processing*, 1952, **38,** June, 288–294, 298 ; July, 340–343). The use of high-chromium ferritic type 430 stainless steel as a substitute for austenitic 18/8 type 302 is considered. The compositions and properties of both are compared, and the forging, heat-treatment, scale removal, and cold press working of type 430 are discussed in some detail. The welding, brazing, and polishing of the ferritic type material, are dealt with in the second instalment. (11 references).—P. M. C.

Properties of Some Spanish Roller Bearing Steels. M. Pujol Roig. (*Inst. Hierro Acero*, 1952, **5,** Aug., 817–821). [In Spanish]. The standard Spanish steel for roller bearing is type 241 with 0·9–1·2% C, 0·1–0·35% Si, 0·3–0·4% Mn, 0·04% S, 0·04% P, 1·4–1·8% Cr and sometimes 0·2–0·3% V. It is supplied oil-quenched from 840–860° C., and tempered at 200° C. giving hardnesses of Rockwell C 64–66 and 60–64, respectively. Two Spanish steels are compared with a Swedish steel. To effect correct martempering, the isothermal transformation curves must be studied for various sizes of specimen so that the production of bainite is prevented.—R. S.

Improvement of Materials Performance in the Mining Industry. B. de Borde. (*Rev. de l'Ind. Min.* 1952, **33,** Aug., 624–631). The main factors leading to the use of alloy steels are listed. The characteristic properties of alloy steels and some special irons are presented in tabular form together with the uses of alloy steels in mining.—A. G.

Överum Steel—An Annealed Cast Iron with Original Characteristics. C. Berg. (*Tekn. Tidskr.*, 1952, **82,** Aug. 12, 639–641). [In Swedish]. Steels developed by the Överum Company, especially for ploughshares, are now finding numerous other applications. They have tensile strengths of 40–140 kg./sq. mm. and elongations of 16–1·5% at 180–400 Brinell. An addition of silicon carbide to this steel is desirable just before pouring, as well as hydrogen degassing during heat-treatment. Ferritic irons are annealed in a tunnel kiln in which the temperature is raised to 950° C. in 25 hr. and kept at this level for 20 hr. Hardening can be in running water directly after annealing, and no microcracking is experienced. An improved method of hardening is also described.—G. G. K.

The Annealing and Heat Treatment of Ductile Cast Iron. C. C. Hodgson and W. Fairhurst. (*Metallurgia*, 1952, **45,** June, 283–288, 296). An account is given of an investigation undertaken to determine the relationship between the thermal treatments applied to ductile cast iron and its structure and mechanical properties. The treatments concerned comprise annealing, for the purpose of improving machinability or developing the maximum ductility and toughness, and hardening and tempering. To obtain optimum impact resistance after annealing, the upper limit of the silicon content should be restricted, and a specification is suggested. Impact figures of ductile cast irons low in manganese after subcritical annealing were comparable with those of blackheart malleable iron. Irons containing about 2% Si gave lower tensile strengths than irons containing about 1½% Si ; owing to decomposition of the martensitic matrix in the former resulting in the random precipitation of a large number of mall graphite particles.—B. G. B.

Designing with Ductile Iron. J. L. Goheen. (*Machine Design*, 1952, **24,** June, 161–166). The characteristics of ductile iron are surveyed and information on important design considerations is given.—E. C. S.

S.G. Cast Irons. (*Automobile Eng.*, 1952, **42,** Sept., 338–340). Some notes are given on the properties and production characteristics of spheroidal graphite cast irons. Typical applications of this material in the automobile industry are listed.—P. M. C.

Strength, Structure, and Composition of Unalloyed Grey Iron. H. T. Angus. (Inst. Brit. Foundrymen : *Foundry Trade J.*, 1952, **93,** Aug. 28, 239–245 ; Sept. 4, 269–273). This is a review, in which the author's emphasis is on practical aspects. He points out the increase of tensile strength, hardness, and transverse strength with the carbon equivalent, and relates the structure to the composition. The cooling rate in relation to the size of casting is discussed.—E. T. L.

Strength of Alloy Cast Irons. H. A. Altorfer. (*Machine Design*, 1952, **24,** Mar., 162–164). An appraisal is made of the limitations of cast steel and cast iron, based on experimental results for cylindrical vessels. It is claimed that these results may be applied to pressure vessels of all shapes.

Acicular Iron Castings. (*Oil Eng. and Gas Turb.*, 1952, **20,** June, 45). An account is given of the application, by the National Gas and Oil Engine Co., Ltd., of this special high-duty cast iron for crankshafts, brake levers, valves, gear wheels, and other parts subject to heavy stresses.—B. G. B.

Boiler Research. (*Elect. Rev.*, 1952, **150,** June 27, 1387–1393). The new research laboratories of Babcock and Wilcox, Ltd., Renfrew, are described.—R. A. R.

METALLOGRAPHY

Microconstituents in High Temperature Alloys. H. J. Beattie, jun. and F. L. Versnyder. (*Amer. Soc. Met. Preprint No.* 1, 1952). Iron-base, cobalt-base, and nickel-base alloys have been studied with the microscope and by X-ray diffraction. The occurrence and probable form of titanium carbonitride in a nickel-base age-hardenable alloy were determined, and are described in detail. The intermetallic compounds M_2Ti and Fe_2Al were identified, and a new phase $CrMoN_Z$ was observed. The second Brillouin zone for the complex M_6C and $M_{23}C_6$ structures has a capacity of about 7 electrons per metal atom.—E. T. L.

Influence of the Rate of Cooling on the Microstructure. E. D. Bia. (*Inst. de Ensayo de Materiales*, 1951, No. 24, *Bol. de la Facultad de Ingenieria*, 1951, **4,** Oct., Reprint). [In Spanish]. The author interprets structures obtained by a Jominy end-quench test on a 0·95% carbon steel.—R. S.

Present Knowledge of the Metallography of Austenitic Alloys Especially in Steels of the 18/8 Type. P. Bastien and J. Dedieu. (*Métaux-Corrosion Indust.*, 1952, **27,** June, 237–344). The evidence, obtained from X-ray diffraction measurements of the presence of carbides of the cementite types, Cr_3C_2, Cr_7C_3, and Cr_4C, is presented in tabular form. The equilibrium diagrams of the system are shown and discussed.—B. G. B.

Observations on Nodular Graphite. H. M. Weld, R. L. Cunningham, and F. W. C. Boswell. (*Trans. Amer. Inst. Min. Met. Eng.: J. Met.*, 1952, **4,** July, 738–742). The authors describe a study of the structure of graphite nodules and of the reason for their formation in nodular cast iron. Metallic iron is randomly entrapped in the nodules and both flake and nodular graphite probably contain a non-ferromagnetic iron compound. There appears to be no element in sufficient quantity to account for a visible nucleus, and the bright central spot seen in polished sections of nodules, is thought to be due to the structure of the nodule. The partition of the magnesium addition between metal and graphite is independent of the form of the graphite.—G. F.

Nodules and Nuclei in Nodular Cast Iron. Part II. J. E. Rehder. (*Amer. Foundryman*, 1952, **21,** May, 87–90). The author reports work on the constitution of nuclei for graphite nodules, using chemical analysis, and X-ray diffraction. He concludes that two patterns of nodular graphite are found—radial and non-radial. The nuclear material is hard and inert, probably with a hexagonal crystal habit. Chemical

analysis of the ash from clean nodules shows that iron, titanium, and aluminium are present, whilst magnesium is there in only small quantities. The graphite nodules show a major decrease in magnetic susceptibility at about 500° C. [For Part I *see J. Iron Steel Inst.*, 1952, **172**, Dec., 456].—E. L.

Carbides in High Speed Steel—Their Nature and Quantity. F. Kayser and M. Cohen. (*Metal Progress*, 1952, **61**, June, 79–85). Four high-tungsten and four molybdenum steels were investigated. Three carbides were found in each of the steels by etching with 2% nital containing Zephiran chloride, namely, M_6C, $M_{23}C_6$, and MC where M represents the sum of the metal atoms. Amounts of the carbides by volume and by weight were determined and compared. The chemistry of the carbides and steel matrix, and the partition of the elements are discussed.—B. G. P.

Anisotropy of the Distortion of Crystal Lattice of Martensite. V. A. Il'ina, V. K. Kristkaya, and G. V. Kurdyumov. (*Doklady Akademii Nauk S.S.S.R.*, 1952, **85**, 5, 997–999). [In Russian]. From X-ray investigations it is concluded that the weakening of double lines 011, 002, and 112 in martensite containing 1·3% of carbon is due to distortion, which seems to be a statistical displacement of atoms from the normal position in the lattice in the direction (001).—V. G.

Statistic Mechanics and the Partition of Numbers. II—The Form of Crystal Surfaces. H. N. V. Temperley. (*Proc. Cambridge Philosophical Soc.*, 1952, **48**, Part 4, Oct., 683–697). The classical theory of partition of numbers is applied to the determination of the equilibrium profile of a simple cubic crystal.—R. A. R.

The Fine Structure of Bragg Reflections and Imperfections of Metallic Crystals. H. Lambot and L. Vassamillet. (*Rev. Mét.*, 1952, **49**, June, 408–410). A method of measuring crystal imperfection is described, in which the specimen is placed at the focal point of a strictly monochromatic X-ray beam. Advantages of the technique are discussed.—A. G.

The Self-Perpetuating Step in Crystal Growth from the Melt. B. Chalmers and Ursula M. Martius. (*Phil. Mag.*, 1952, **43**, June, 686–687). The authors discuss the mechanism of the nucleation of new layers on a crystal growing from the melt. It is suggested that atoms joining the solid attach themselves in the re-entrant corners of the corrugated solid–liquid interface. The effect of this is to move the corrugations with the temperature gradient, without reducing the re-entrant sites.

Precipitation from Solid Solution. J. C. Fisher, J. H. Holloman and J. G. Leschen. (*Indust. Eng. Chem.*, 1952, **44**, June, 1324–1327). Precipitation in solids differs from that in liquids and vapours mainly by coherency relationships between the parent and precipitating phases. The free energy of coherent interfaces is small, and coherent precipitates form whenever the transformation free-energy change is sufficient to provide the elastic coherency strain. Observations on precipitates formed in various temperature ranges and mechanical property changes can be understood in terms of coherent and non-coherent precipitation. (12 references).

Nucleation Catalysis. D. Turnbull and B. Vonnegut. (*Indust. Eng. Chem.*, 1952, **44**, June, 1292–1298). Structures of crystals and substances which catalyse their formation closely resemble one another on certain atomic planes. A crystallographic theory of nucleation catalysis predicts that catalytic potency should be related with the degree of disregistry between catalyst and crystal on certain planes; according to conditions, nuclei can form coherently with the catalyst or by regions of good fit separated by a dislocation gridwork. Interface energies can be predicted. The theory is supported by experimental evidence, and quantitative limits for coherent nucleus formation are established. (32 references).—K. E. J.

Theory of Nucleation in Solids. R. Smoluchowski. (*Indust. Eng. Chem.*, 1952, **44**, June, 1321–1323). The basic problem of a theory of nucleation in solids is outlined, and complications caused by departure from the conditions for applicability of Volmer's condensation theory are described. Existing theories and comparisons with experimental data are summarized; emphasis is placed on the nucleation theory of crystallization and its successful interpretation of observations. The qualitative agreement with experiment confirms general ideas concerning nucleation, but the absence of quantitative agreement does not permit a critical appraisal of the specific assumptions. (20 references).—K. E. J.

Nucleation in Phase Transitions. V. K. La Mer. (*Indust. Eng. Chem.*, 1952, **44**, June, 1270–1277). An understanding of nucleation processes is necessary in the preparation of all types of colloidal dispersions by condensation methods; *e.g.*, fog or rain production by seeding; initiation and production of new metallurgical phases that alter physical properties; or chemical supercooling, superheating, overcompression, supersaturation, and precipitation. A critical historical review is given of theoretical and phenomenological aspects of nucleation. (33 references).—K. E. J.

Statistical Theory of Properties of Solid Solutions. R. Smoluchowsky. (*Phys. Rev.*, 1951, **84**, Nov. 1, 511–518). Properties of binary solid solutions are considered in terms of local fluctuations in lattice composition. These influence alloy properties by varying the corresponding local concentration of electrons. A simple statistical method is given for calculating properties of random and ordered solid solutions. The theory is applied to saturation magnetization, temperature coefficient of electrical resistivity, thermoelectric power, and other properties of various iron and other alloys, showing satisfactory agreement with experimental data. (16 references).

The Mechanical Properties of Solids as a Function of Temperature. P. Le Rolland. (*Métaux-Corrosion-Indust.*, 1952, **27**, May, 216–232). The atomic structure of materials is first discussed together with the properties and structure of single perfect crystals. The properties and structure of actual polycrystalline materials are next reviewed and the phenomenon of hardening is related to the structural changes. The rôle of internal tensions, intergranular joints, and the influence of foreign elements and internal friction on the properties of polycrystalline materials are discussed.—B. G. B.

Effect of Carbon Content on 18–4–1 High-Speed Steel. A. H. Grobe and G. A. Roberts. (*Amer. Soc. Met. Preprint No.* 15, 1952). Iso-hardness curves, relating austenitizing temperature and time to the quenched hardness, are presented for six 18–4–1 W–Cr–V steels, of various carbon contents between 0·52 and 0·78%. Metallographic studies have shown the presence of ferrite at high austenitizing temperatures in the lower carbon steels, which is not evident at lower austenitizing temperatures. A chart is provided which describes the tempering cycle to give any desired hardness above Rockwell C-45. The impact strength has been studied as a function of carbon content, austenitizing temperature, and hardness.—E. T. L.

Sigma Formation and Its Effect on the Impact Properties of Iron–Nickel–Chromium Alloys. A. M. Talbot and D. E. Furman. (*Amer. Soc. Met. Preprint No.* 2, 1952). The approximate sigma-phase boundaries between 650° and 900° C. have been located for Fe–Ni–Cr alloys high in nickel and chromium. For conditions near equilibrium, the boundary extends from about 21% Cr at 20% Ni, to 24% Cr at 35% Ni. It is shown that silicon moves the sigma area to lower chromium contents. Embrittlement, as measured by impact tests at room temperature, increased rapidly with the first few per cent of sigma formed, regardless of the composition of the base material.—E. T. L.

The Effect of Composition on the Temperature of Spontaneous Transformation of Austenite to Martensite in 18/8 Type Stainless Steel. G. H. Eichelman, jun., and F. C. Hull. (*Amer. Soc. Met, Preprint No.* 9, 1952). Dilatometric measurements have been used to find the relative efficiency of various additions in lowering the transformation temperature. The efficiency increases in the following order: Silicon, manganese, chromium, nickel, carbon, nitrogen. Changes in the transformation temperature caused by carbide precipitation can be used to determine the solubility of carbon in stainless steel.—E. T. L.

The Effect of Silicon on the Tempering of Martensite. A. G. Allten and P. Payson. (*Amer. Soc. Met. Preprint No.* 10, 1952). Hardness, dilatometric, density, electrical resistance, and X-ray diffraction measurements demonstrate that the temperature at which cementite forms, on heating martensite at a definite rate, rises with increasing silicon in the steel. The temperature at which cementite formed in the 3% Ni steels, was raised about 150° C. by an increase of silicon from 0·5 to 2·2%. Epsilon iron carbide was detected in the 0·6% C. steels after tempering at lower temperatures than those required to produce cementite. The hardness after tempering at between 205° and 315° C. was found to depend only on the carbon and silicon contents of the steel. An increase in silicon from 0·46% to 2·18% in a 3% Ni steel raised the tempering temperature for minimum impact strength from 600° to 800° F.—R. A. R.

Effects of High-Temperature Heating on the Isothermal Formation of Bainite. S. A. Cottrell and T. Ko. (*J. Iron Steel Inst.*, 1953, **173**, Mar., 224–228). [This issue].

The Electrolytic Separation and Some Properties of Austenite and Sigma in 18–8–3–1 Chromium–Nickel–Molybdenum–Titanium steel. T. P. Hoar and K. W. J. Bowen. (*Amer. Soc. Met. Preprint No.* 4, 1952). The austenite and sigma phases are extracted from the steel by heat-treatment at 850° C., followed by selective anodic dissolution in 25 wt-% sulphuric acid at 0·4 amp./sq. in. and in 50 vol–% hydrochloric acid at 30 amp./sq. in., respectively. The extracted austenite partially transforms at room temperature, but this is halted at −196° C. The transformation of extracted austenite is thus easier than that of massive austenite. The extracted sigma is ferromagnetic at low temperatures. It is much higher in chromium and molybdenum, and to a lesser extent in titanium, than the corresponding austenite. Electrode potential measurements of compacted austenite and sigma residues in sodium chloride and hydrochloric acid solutions show that sigma has a greater tendency to become passive, and has a more noble passive potential than has the austenite.—E. T. L.

Note on the Determination of TTT Curves. M. Casadio and M. Ferrero. (*Met. Ital.*, 1952, **44**, Aug.–Sept., 354–360). A description is given of the method adopted and the equipment used to obtain data for determining **S**-curves. The difficulties encountered in determining these curves for certain steels at temperatures between 500° and 650° C. are discussed in detail. Mention is made of new apparatus with continuous recording which should enable quantitative measures to be made of some of the variables.—M. D. J. B.

The Application of X-Ray Spectrometer with Geiger Counter for the Direct Determination of the Velocities of Structural Transformations. U. Rösler and W. Ruff. (*Met. Ital.*, 1952, **44**, Aug.–Sept., 475–487). A new experimental and mathematical method is described for the X-ray study of the kinetics of transformations at high temperatures, employing a Philips X-ray spectrometer with Geiger counter. Factors influencing the experimental results are discussed. The application of a new mathematical method of analysing the intensity–time curves obtained makes it possible to establish, at every instant, the speed of phase formation and deformation which occur in structural transformations, even if the phases are unstable and short-lived at high temperatures. In order to demonstrate the method, the authors examined a bronze–aluminium alloy. (27 references).—M. D. J. B.

Modern Metallurgy of Cast Iron. H. Morrogh. (*J. West of Scotland Iron Steel Inst.*, 1949–50, **57**, 63–80). Cast iron is considered as a ternary or quaternary alloy. White and grey structure are discussed with particular reference to the method of formation from the liquid state. Eutectic structures, fine graphite and flake graphite are mentioned, and nodular graphite is dealt with more fully.—P. C. P.

CORROSION

Corrosion Problems in Industry. W. H. J. Vernon. (*Chem. Ind.*, 1952, June 14, 524–526). Such sources of trouble as the presence of anodic areas, small in relation to the cathodic metal, dissimilar metals, excessive or insufficient aeration, and faulty design are clearly recognized and should be avoided wherever natural water, solutions, or condensates are handled. The choice of materials, coatings, or surface finish is an important preventive factor. Adequate recognition of the fundamental requirements for the virtual elimination of corrosion failure can make a valuable economic contribution to industrial progress.—J. O. L.

Corrosion of Metals. U. R. Evans. (*Chem. Ind.*, 1952, Oct. 11, 986–992). This historical survey of the subject treats chronologically with the growth of interest in and understanding of the mechanism of corrosion. Although its effects may have been observed earlier, the increasing use of sulphur-bearing fuels aggravated its occurrence and in the 19th century, study of the electrochemical mechanism began. Theories based on chemical reaction were posed and discarded, bimetal contacts studied, and oxidation, atmospheric attack, corrosion fatigue, and stress corrosion were investigated. These researches have led to the increasingly well-informed use of inhibitors, cathodic protection, and surface coating, all of which had previously been employed on an empirical basis.

The Fight Against Rust. (*Brit. Iron Steel Res. Assoc. Pamphlet*, Oct., 1952). This 25-page booklet gives an account of the work of the Corrosion Committee of B.I.S.R.A., and of the results of this work that can be useful to industry.

Corrosion Testing. S. C. Britton. (*Research*, 1952, **5**, Aug., 377–384). A survey is made of service, field, and laboratory testing of corrosion behaviour. Research has given valuable knowledge of the mechanism and velocity of corrosion, the build-up of corrosion layers, effects of trace elements, surface conditions, and damage to structure. Testing procedures are described for immersed corrosion, atmospheric corrosion, buried metals, and acceptance of prepared or protected specimens. Some results are given for zinc, ingot iron, and copper-bearing steels. (46 references).—K. E. J.

Cathodic Protection of Steel. F. Wormwell, T. J. Nurse and H. C. K. Ison. (*Chem. Ind.*, 1952, Oct. 4, 972–974). The electrical requirements for the protection of steel in sea water by means of sacrificial zinc anodes or by an applied e.m.f. are considered, also the use of the latter method to inhibit bacterial activity, which leads to attack in suitable media. An e.m.f. of about −1·0 V. was found to be adequate under the conditions investigated, but it is shown that both current density and potential measurements are required to define the protective action in different environments.—J. O. L.

Cathodic Protection of Steel in Contact with Water. L. P. Sudrabin and H. C. Marks. (*Indust. Eng. Chem.*, 1952, **44**, Aug., 1786–1791). Current requirements for the cathodic protection of submerged steel are controlled by the same environmental factors which influence corrosion rate ; they are increased by higher dissolved oxygen concentrations, temperature, or velocities which increase oxygen diffusion rate, but are decreased by deposited calcium carbonate and resistant coatings. Where protective paints and cathodic protection are combined, best results are obtained by control of the potential at the paint surface to eliminate activity at the flaws and damage to the coatings. (30 references).

The Corrosion of Gas-Burning Appliances. J. Welsh. (*Chem. Ind.*, 1952, June 21, 551–555). The residual sulphur, particularly as organic compounds, present in town gas, even after purification, is sufficient to cause serious corrosion of gas appliances. Measurements of the H_2SO_4 content, and reactivity of combustion products, has led to the development of fresh lines of approach to the problem of control, which may well need to be associated with modifications in design and materials to achieve maximum resistance to attack.—J. O. L.

Control Galvanic Corrosion—Stop Costly Failures. E. C. Reichard. (*Mat. Methods*, 1952, **36**, Aug., 83–86). A brief outline of the mechanism of galvanic corrosion is given. A galvanic series of the commoner engineering metals is shown, and several methods of eliminating or controlling the destructive effects of galvanic action are discussed. These include protective coatings, inhibitors, and sacrificial anodes.—P. M. C.

Corrosion Resistance of Titanium, Zirconium, and Stainless Steel. L. B. Golden, I. R. Lane, jun., and W. L. Acherman. (*Indust. Eng. Chem.*, 1952, **44**, Aug., 1930–1939). Comprehensive results are given of corrosion-rate tests (using A.S.T.M. recommended apparatus) on titanium, zirconium, and 20/29 chromium–nickel stainless steel exposed to sulphuric, hydrochloric, nitric, phosphoric, and mixed acids ; water-saturated chlorine ; ferric, cupric, mercuric, stannic, nickel, manganous, sodium, ammonium, calcium, magnesium, barium, zinc, and aluminium chloride solutions, and hypochlorite solutions. (22 references).—K. E. J.

Corrosion. M. G. Fontana. (*Indust. Eng. Chem.*, 1952, **44**, Aug., 87A–90A). Knife-line attack (*i.e.*, intergranular corrosion adjacent to welds) has been investigated for various grades of stainless steel stabilized with niobium, exposed to nitric acid under varying conditions. These steels were previously considered immune from intergranular corrosion. Those with low carbon content require longer at sensitizing temperature to develop this attack.—K. E. J.

Protection of Metals against Pitting, Tuberculation, and General Corrosion. H. L. Kahler and P. J. Gaughan. (*Indust. Eng. Chem.*, 1952, **44**, Aug., 1770–1774). The use of chromate and phosphate inhibitors of corrosion is discussed. Laboratory and plant results show that a combined phosphate–chromate inhibitor, used under various conditions and in low concentrations, was more effective than either single treatment for controlling pitting, tuberculation, and general corrosion. The combination acted as a single mechanism, and surface alteration amounted to slight etching only.—K. E. J.

Effect of Temperature on Corrosion of Metals by Water. N. Hackerman. (*Indust. Eng. Chem.*, 1952, **44**, Aug., 1752–

1755). The influences of systematic temperature gradients on rates and distribution of corrosion reactions are considered. A general treatment is impossible, owing to the variety of the influences of temperature on the components of corrosion systems, but some explanations are possible through systematic analysis. Some major temperature-dependent properties discussed are oxygen solubility, galvanic potential, and precipitation rate. (21 references).—K. E. J.

Effects of Velocity on Corrosion by Water. H. R. Copson. (*Indust. Eng. Chem.*, 1952, **44,** Aug., 1745–1752). This is a general review of the effects of velocity on corrosion by water, with emphasis on iron, zinc, copper and their alloys : some unpublished data are included. Corrosion generally increases with velocity, but the opposite effect is possible. Motion may eliminate, or initiate, local attack, and can greatly affect galvanic couples. At high velocities, mechanical damage accelerates attack. Each problem requires full operating data, and consideration of secondary effects. (56 references).

Action of Sodium Silicate as a Corrosion Inhibitor in Water Piping. L. Lehrman and H. L. Shuldener. (*Indust. Eng. Chem.*, 1952, **44,** Aug., 1765–1769). Results from several hundred buildings in many cities show that a very effective treatment is the continuous addition of sodium silicate. A non-growing film, mainly silica, forms on the pipe : the mechanism is not clearly understood. Equilibrium between ionic and colloidal SiO_2 was demonstrated. Silica is removed from the water on to zinc, copper, or iron corrosion products : with zinc, chemisorption was established. (24 references).

Cooling Water Problems in the New York Metropolitan Area. S. Sussman. (*Indust. Eng. Chem.*, 1952, **44,** Aug., 1740–1744). In the New York area, soft, low-alkalinity make-up water, and heavy atmospheric contamination create severe corrosion problems. CO_2 and SO_2 can reduce the *p*H of water below 3·5. Best control is provided by alkali for the water, and chromates or silicates for passive film formation. Poor equipment or maintenance can counteract preventive measures. Where ground water is used, scale formation, organic growths, and accelerated corrosion are problems.

Corrosion Control in Water Systems. H. H. Uhlig. (*Indust. Eng. Chem.*, 1952, **44,** Aug., 1736–1740). This is a general survey of means of mitigating corrosion, covering cathodic protection, metallic coatings, organic coatings, inhibitors, water treatment, and container composition. Of means available to prevent deterioration, economic and technical factors indicate that cathodic protection, paints, and organic coatings will increase in applications. Of corrosion-resistant metals, titanium is most promising, for use in wrought and cast form, or as a clad coating. (18 references).—K. E. J.

Inhibitors for Eliminating Corrosion in Steam and Condensate Lines. R. C. Ulmer and J. W. Wood. (*Indust. Eng. Chem.*, 1952, **44,** Aug., 1761–1765). Test data from 600 lb./sq. in. boiler condensate showed that introduction of alkali was more effective than the use of coating-forming materials in the case of iron : both types of treatment gave good results with copper and brass. Use of the two methods in different parts of the system is possible. With both types, rate of corrosion increases at higher concentrations of dissolved oxygen. (22 references).—K. E. J.

Corrosion Control with Organic Inhibitors. J. N. Breston. (*Indust. Eng. Chem.*, 1952, **44,** Aug., 1755–1761). In many large-scale operations involving water, inorganic corrosion inhibitors have little effect. Some organic compounds, however, offer protection by physical and chemical adsorption on corroding surfaces. They form films of molecular thicknesses, and are effective in extremely low concentrations. The mechanisms and properties of organic inhibitors are described : Applications are outlined for acid cleaning solutions, cooling systems, steam systems, oil wells, and petroleum-handling plant. (57 references).—K. E. J.

Sodium Benzoate as a Corrosion Inhibitor. (*Product Finishing*, 1952, **5,** Aug., 90–97). The corrosion-inhibiting properties of sodium benzoate in aqueous solution or when used as an impregnating agent for paper, regenerated cellulose, rubber latex, and other film materials is described.—J. P.

Research on Corrosion in Salt Sprays. G. Bianchi. (*Met. Ital.*, 1952, **44,** Aug.–Sept., 330–335). The author studied the weight lost by corrosion of iron and zinc test pieces submitted to salt sprays. The influence which the methods of spraying, the duration of test and the position of the sample, have on the ultimate corrosion are examined. The author indicates the best experimental techniques for salt-spray corrosion tests and explains the process of corrosion.

A Critical Study of Salt Spray Tests. F. L. LaQue. (*Métaux-Corrosion–Indust.*, 1952, **27,** May, 233–236). The use of a salt spray for testing metallic specimens for resistance to corrosion is reviewed. Methods of measuring the efficiency of metallic and non-metallic coatings, the thickness of the coating and the detection of flaws in the coating are described. A comparison of the corrosion resistance of different metals and alloys is made.—B. G. B.

Protection of Stainless Steels from Pitting in Salt Solutions. L. Cavallaro and C. Bighi. (*Met. Ital.*, 1952, **44,** Aug.–Sept., 361–365). The authors studied the inhibiting effect of small additions of caustic soda to saline solutions in which specimens of 18/8 steel and 18% Cr steel were immersed. It is shown that pitting, normally prevented by additions of molybdenum to the metal, may be prevented by very small (0·1%) additions of caustic soda to the solution.—M. D. J. B.

Highly Alloyed Stainless Has More Acid Resistance. H. Thielsch and W. E. Pratt. (*Iron Age*, 1952, **170,** July 10, 135–139). Stainless steels highly alloyed with copper, molybdenum, nickel, and chromium show excellent corrosion resistance to a wide range of sulphuric acid concentrations. These alloys can be welded. Tests show presence of sigma and the copper-rich phases may seriously reduce corrosion resistance to highly active solutions. Control of composition and annealing at temperatures between 1900° and 2100° F. are recommended.—A. M. F.

Studies in the Corrosion of Metals Occasioned by Aqueous Solutions of Some Surface-Active Agents. III—Tin. T. K. Ross. (*J. Appl. Chem.*, 1952, **2,** Sept., 526–531).

Cavitation Mechanics and Its Relation to the Design of Hydraulic Equipment. R. T. Knapp. (*Inst. Mech. Eng. Advance Copy*, Apr. 18, 1952).

Corrosion in Small Boilers. J. C. Stainton. (*J. Inst. Heating Ventilating Eng.*, 1952, **20,** Nov., 313–324). The electrolytic, or ionic, theory of corrosion is discussed, particularly in relation to the corrosion of iron, and the conditions under which corrosion can take place are summarized.–R. A. R.

The Problem of Intercrystalline Corrosion of Ferritic, and Predominantly Ferritic, Chromium Steels. E. Houdremont and W. Tofaute. (*Inst. Hierro Acero*, 1952, **5,** Apr.–June, 628–641). [In Spanish]. This is a Spanish version of a paper which appeared in *Stahl u. Eisen*, 1952, **72,** May 8, 539–545 (see *J. Iron Steel Inst.*, 1952, **172,** Nov., 362).

The Passivity of Metals. Part XI—The Anodic Behaviour of Iron Oxide Films. U. R. Evans and I. D. G. Berwick. (*J. Chem. Soc.*, 1952, 3432–3437). The behaviour of iron, coated with visible oxide film by previous heating in air when subjected to anodic treatment in dilute H_2SO_4, has been studied. The time needed for the destruction of the oxide film by autoreduction after interruption of the current was measured. This breakdown time (t_b) increases with increase of the time during which the current flows before the interruption. The increase in t_b is attributed to the increase in the oxygen built up by the anodic polarization. Similar results were obtained at different current densities and at different temperatures.

Sodium Benzoate and Other Metal Benzoates as Corrosion-Inhibitors in Water and in Aqueous Solutions. F. Wormwell and A. D. Mercer. (*J. Appl. Chem.*, 1952, **2,** Mar., 150–160). Sodium benzoate is shown to be an effective inhibitor of the corrosion of mild steel in distilled water, a moderately hard mains water, and very dilute (*e.g.*, 0·03%) NaCl solutions. The concentration of benzoate required for inhibition is greater (0·5%) for machined than for emeried surfaces (0·1% in favourable conditions) and for mains water or chloride solutions (1·0 or 1·5%) as compared with distilled water (0·5%). Sodium benzoate is less efficient than sodium chromate ; it is, however, a safe inhibitor, since it does not lead to intense local corrosion when the concentration is just below the minimum required for protection. The following benzoates also possess inhibitive properties : Potassium, lithium, zinc, and magnesium.—E. C. S.

A Contribution to the Study of the Action of Certain Inhibitors upon Corrosion in Chloride-Polluted Water. W. S. Patterson and A. W. Jones. (*J. Appl. Chem.*, 1952, **2,** May, 273–280). The comparative effects of potassium dichromate, sodium nitrite, sodium benzoate, and soluble oil as corrosion inhibitors were partially examined. The corrosion medium was either diluted sea water or distilled water containing 0·75% (w/v) added sodium chloride, *i.e.*, corresponding to

approx. 25% of sea water addition with respect to chloride content. Steel is readily protected in the presence of chlorides, but cast iron gives somewhat anomalous results. Soluble oil suppresses corrosive attack.—E. C. S.

Prevention of Corrosion by Means of Inhibitors. R. S. Thornhill. (*Research*, 1952, **5**, July, 324–332). After a review of the types of corrosion that may be encountered, and the basic action an inhibitor performs, a number of specific inhibitors in each category is dealt with. Each one of them has some limitation, and its suitability can only be gauged from full-scale or pilot-plant trials. Research, at present, is mainly centred on mixtures of inhibitors to secure non-fitting types at lower cost but greater efficiency.—J. C. B.

Physical Chemical Aspects of Corrosion Inhibition. N. Hackerman. (*Corrosion*, 1952, **4**, Apr., 143–149). Physical and chemical phenomena that can occur during the inhibition of the corrosion of metals by liquid environments are discussed. The way in which each of these may be involved and its relative importance in the overall process are considered. Several recent researches in this field are examined.—G. P. A.

Crystallochemical Mechanism of Formation of Oxide Film on Iron at Room Temperature. P. D. Dankov and N. A. Shishakov. (*Zhurnal Fizicheskoi Khimii*, 1949, **23**, 9, 1031–1035). On the basis of X-ray diffraction measurements, a mechanism for the formation of iron oxide films on iron is postulated.—V. G.

Chemical Composition and Corrosion in Relation to the Underground Mine Waters of the Witwatersrand. F. C. Johnson. (*J.S. African Inst. Mech. Eng.*, 1952, **1**, May, 285–315). Underground corrosion troubles chiefly affect water pipes, pumping columns, shaft guides, haulage tracks, and ore wagons, but corrosive water in the presence of humid air also does great damage to steel structures. No criterion of corrosiveness applicable to mine water has yet been established, but methods of protection are outlined. Soluble oil is an effective protection against internal corrosion. Calcium sulphate is added to acid waters, and methods of restoring saturated conditions in waters which have been supersaturated with thin salt are discussed.—H. R. M.

A Hydrogen Evolution Method for Evaluation of Corrosion Inhibitors for Oil Wells. W. R. Scott and G. H. Rohrback. (*Corrosion*, 1952, **8**, July, 234–240). Oil-well corrosion is an electrochemical reaction, hydrogen ions for the cathodic reaction being produced by dissociation of carbonic acid, of organic acids, and, in many wells, of hydrogen sulphide. Inhibitors provide one of the most promising methods for controlling this corrosion. A method of evaluating inhibitors, based on measurement of hydrogen evolution, is described, and details of the apparatus and results for various inhibitors are given.—G. P. A.

Performance of Magnesium Galvanic Anodes in Underground Service. O. Osborn and H. A. Robinson. (*Corrosion*, 1952, **4**, Apr., 114–129). The performance of magnesium anodes in various backfills and in different soils has been studied for five years in field tests. The effects of anode composition, time, backfill composition, current density, and soil characteristics are discussed. Impurities in the anodes must be kept low for good anode efficiency. A high purity magnesium alloy (with 6% Al and 3% Zn) has given the best performance. At current flows equivalent to a 10-year life, efficiencies corresponding to 500 amp.-hr./lb. were obtained, except in high-chloride soils where a higher current density was needed for comparable efficiency. Efficiency of anodes increases with current density and with time for periods up to at least two years. Magnesium sulphite and chromic acid were the only backfills found to have a significent effect on efficiency. Backfills installed dry give higher current output than those that are installed wet. With good drainage, soluble parts of the backfill are leached out, but current flow is not affected as long as moisture is adequate.—G. P. A.

Use of Magnesium Anodes in Cathodic Protection of Tanks and Pipelines. L. E. Magoffin. (*J. Amer. Water Works Assoc.*, 1952, **44**, May, 407–412). The protection of two pipelines and a water tank is described. The calculations of the number of anodes required are shown. The results obtained have been very satisfactory. Although the overall cost of magnesium anodes for protection has proved slightly higher than that of rectifier-type cathodic protection, its ease of handling and adaptation to difficult situations has more than offset the additional cost.—B. G. B.

Corrosion and Cathodic Protection of Pipelines. W. R. Schneider. (*J. Amer. Water Works Assoc.*, 1952, **44**, May, 413–427.) A summary of part of the existing knowledge of the subject is given. Soil-resistance tests and methods of measurement of pipeline currents are described. The principles of various methods of cathodic protection are discussed. A number of graphs for use in the application of horizontal and vertical electrodes for protection are given. The use of graphite and magnesium anodes is also discussed.—B. G. B.

Vigilance Underground—A Summary of Progress on Preventing Electrochemical Corrosion on Buried Pipe Lines. B. Patterson. (*Corrosion*, 1952, **8**, July, 241–246). The need for control of corrosion on buried metal is explained and elements of planning electrochemical control are outlined. A survey of specifications and inspection fundamentals is followed by recommendations concerning electronic inspection after pipelines are in service.—G. P. A.

Cathodic Protection of Steel Water Tanks. C. K. Wells. (*J. Amer. Water Works Assoc.*, 1952, **44**, May, 428–434). The design of an anode system to distribute a uniform current density of 0·3 milliamp./sq. ft. to all submerged surfaces is described ; twenty anodes of 26-gauge platinum were distributed throughout the tank. The electricity consumption was 20 V. at 10 amp. After cathodic protection had been in operation for seven months the tank was emptied and the installation had given complete protection.—B. G. B.

The Tucumcari Tank Failure. C. E. Barnhart, M. C. May, D. R. W. Wager-Smith, and A. P. Bailey. (*J. Amer. Water Works Assoc.*, 1952, **44**, May, 435–441). A circular steel tank 115 ft. in dia. and 30 ft. high containing 2,300,000 gallons of water, failed. The tank was originally designed for oil storage and was taken down and re-erected as a water tank. A poorly welded joint between two $\frac{1}{2}$-in. steel plates failed throughout its vertical length of 7 ft. The tank was not designed for use with water and should not have been so used.—B. G. B.

Corrosion Problems of the Railroads. L. S. Crane. (*Corrosion*, 1952, **4**, Apr., 149–151). Corrosion costs on American railways are estimated at 268 million dollars annually. The types of corrosion affecting the various items of equipment are described, and some major problems are cited.—G. P. A.

Some Observations of the Potentials of Stainless Steels in Flowing Sea Water. K. M. Huston and R. B. Teel. (*Corrosion*, 1952, **8**, July, 251–256). Potentials of 12 stainless steels in turbulent sea water have been investigated at the Harbor Island Marine Test Station. Potential is related to the composition of the stainless steel, the straight chromium grades giving the highest average figures. Sulphur, the free-machining element, increases the average potential. Higher potentials are associated with greater susceptibility to crevice corrosion.—G. P. A.

Special Tests for Stainless Steel. J. Lomas. (*Brit. Steelmaker*, 1952, **18**, Aug., 414–417). The author describes a number of tests applied to stainless steels to determine the susceptibility to corrosion and to assess its extent, and he discusses their comparative values. Methods of studying intercrystalline corrosion are also considered.—G. F.

ANALYSIS

Instrumentation. R. H. Munch. (*Indust. Eng. Chem.*, 1952, **44**, Aug., 81A–83A). Details are given of the operating principles of infra-red absorption spectrometers for laboratory analysis of complex mixtures of compounds, and of infra-red photometers for continuous analysis of gas streams on the plant (non-selective detector and selective receiver types).

A Laboratory Mass Spectrograph with Many Uses. G. Philbert. (*J. Phys. Radium*, 1952, **13**, Feb., 50A–58A). In the apparatus described, the gas pressure is as high as 10^{-2} mm. Hg, in the ionization range. It allows the possibility of altering the ion source, and provides for the study of visible ionization. The resolution of the instrument is as good as that of machines produced commercially. Full details are given of the construction and electrical circuits.—E. T. L.

Metallurgical Applications of X-Ray Fluorescent Analysis. P. K. Koh and B. Caugherty. (*J. Appl. Phys.*, 1952, **23**, Apr., 427–433). Techniques are described for fluorescent analysis using an X-ray Geiger counter goniometer. The observed intensity of fluorescent radiation from several common metallic elements, including iron, was compared under a constant primary radiation that can be closely maintained in

practice. Intensities of fluorescent spectra were measured on constituents of Fe–Ni, Fe–Cr, and Fe–Ni–Cr alloys.—R. A. R.

Behaviour of Nitrogen and Some of Its Compounds in Steel. H. F. Beeghly. (*Analy. Chem.*, **24**, July, 1095–1100). The importance of determining how much nitrogen is combined with a given element in steel is explained. A micro-Kjeldahl method for 'soluble' nitrogen and a modified procedure for 'insoluble' nitrogen are outlined. The use of an ester-halogen reagent is mentioned, and the preparation of ammonia-free water using ion-exchange resins is described. Some results using these methods are given.—T. E. D.

Determination of Nitrogen and Sulphur in Steel. (*Analy. Chem.*, 1952, **24**, Jan., 199–205). This is a report of a round-table discussion held by the Division of Analytical Chemistry at the 119th Meeting, Amer. Chem. Soc., Cleveland, Ohio, Apr., 1951. A survey is made of methods for the determination of nitrogen, including that in various alloy steels. Detailed procedures, using macro-and micro-Kjeldahl techniques, are given, but it is stated that the method used must be adapted for any particular sample. Three sulphur determination methods were considered : Gravimetric, evolution of H_2S, and combustion in oxygen. The last, though empirical, was said to give the most reliable results. Exact procedures for both carbon steels and alloy steels are given, and various absorbents and titration techniques are discussed.—T. E. D.

Permanent Recording System for Single-Sweep Cathode-Ray Polarographs. F. C. Snowden and H. T. Page. (*Analy. Chem.*, 1952, **24**, July, 1152–1156). The circuit for using a high-speed recording galvanometer instead of photographic recording is described. The rapid response, and the chart speed of 10 cm./sec. give better resolution of the polarogram ; this is especially useful when investigating the kinetics of rapid reactions.—T. E. D.

Contribution on the Determination of Nitrogen in Certain Alloy Steels and Ferro-Alloys. P. Rocquet and M. Olette. (*Rev. Mét.*, 1952, **49**, Mar., 221–230). A method capable of determining nitrogen in alloys containing insoluble nitrides has been developed in which the metal is treated *in vacuo* with sodium. The volume of gas evolved is measured by a new gas analyser using dry absorbents. Only a few grammes of the sample are needed.—A. G.

Some Statistical Considerations Regarding the Chemical Analysis of Cast Iron. J. Gelain. (*Inst. Hierro Acero*, 1952, **5**, Apr., 417–425). [In Spanish]. The French Centre Technique des Industries de la Fonderie periodically organizes 'comparison of analyses,' in which more than 80 members participate. Statistical methods are applied to the results of individual laboratories as well as to the group of analyses. The results of recent analyses for carbon show a distribution approximating to normal in the sense of the Laplace and Gauss laws. Useful conclusions are drawn regarding the reliability of the determinations.—R. S.

Analysis of Chromium Plating Baths Containing Fluorides and Fluosilicates. H. A. van Oosterhout and E. J. C. van de Velde. (*Metal Finishing*, 1952, **50**, Apr., 60–63, 69). The use of the distillation–alizarin–sulphonate titration method, for determining fluoride and fluosilicate in chromium plating baths is described. When both fluoride and fluosilicate are present, the total fluoride is estimated in this way and the fluosilicate by weighing precipitated silica.—J. P.

Solid and Gaseous Fuels. H. T. Darby. (*Analy. Chem.*, 1952, **24**, Feb., 244–251). The analyses of solid, liquid, and gaseous fuels are comprehensively reviewed, with particular reference to recent methods and techniques. (129 references).

Determination of the Metallic Iron in Slags. E. Stengel. (*Technische Mitteilungen Krupp, Forschungsberichte*, 1943, **6**, Mar., No. 8, 137–141). The mercurous chloride method for the determination of metallic iron is not suitable for use with basic slags. By replacing the usual solvent (water) by alcohol, this method can be applied for certain basic slags. Numerical results are given for various types of slag together with details of the procedure.—R. J. W.

Application of the High Frequency Titration. I—Rapid Determination of Lime in Basic Slags. S. Musha. (*Sci. Rep. Res. Inst. Tôhoku Univ.*, 1951, **3A**, Feb., 55–61). A high-frequency titrimeter that measures plate-current changes during titration is described.—J. G. W.

The Direct Determination of Alumina in Silica and Alumino-Silicate Refractories. (*Trans. Brit. Cer. Soc.*, 1952, **51**, Sept., 438–487). The most effective method for the decomposition of firebrick and sillimanite samples is to sinter at 1200° C. with a mixture of Na_2CO_3 and $Na_2B_4O_7$ for 15 min. Initial treatment with HF, to remove interfering amounts of SiO_2, is essential for quartzite and silica brick samples. The decomposition is completed by fusion at 1000° C. with a mixture of Na_2CO_3 and $Na_2B_4O_7$. In these materials, Al_2O_3 has been determined directly by precipitation of the 8-hydroxyquinoline complex from a buffered acetate solution. Most of the iron is eliminated from the solution by treatment with NaOH or by filtration of the aqueous extracts of Na_2CO_3 fusions. The remaining iron is converted to the ferrous-0-phenanthroline complex, which is not precipitated by 8-hydroxyquinoline. The Al complex is precipitated at 40°–50° C., then warmed to 70° C., and digested 10 min. to make the precipitate granular. After a standing period, it is filtered and the determination is completed either gravimetrically, or volumetrically.—E. C. S.

Determination of Calcium and Magnesium in Limestones and Dolomites. J. J. Banewicz and C. T. Kenner. (*Analy. Chem.*, 1952, **24**, July, 1186–1187). The analytical procedure for the estimation of calcium and magnesium using ethylene-diamine-tetra-acetic acid (Versene) as complexing agent is described in detail, including the method of separation of calcium by ammoniacal buffered oxalate precipitation.—T. E. D.

Trends in the Design of Laboratory Apparatus. A. O. Beckman. (Scientific Apparatus Makers Assoc.: *Ing. Vetenskaps Akad. Handl.*, 1952, **23**, 2, 92–97). [In English]. After outlining factors that influence design and development of laboratory apparatus, the author discusses discernible trends due to such factors. Trends noted include the merging of laboratory and industrial equipment, continuous quality control and automatic operation of electronic instruments, and the tendency to use miniature instruments.—G. G. K.

Boliden's Central Laboratory. S. Walldén and S. Kjerr. (*Ing. Vetenskaps Akad. Handl.*, 1952, **23**, 1, 17–31). [In Swedish]. Plans are shown of this new laboratory, completed in 1948 at a cost of £220,000. Organization and work of the metallurgical and other departments are described and the construction of the building is discussed.—G. G. K.

New Klip Chemical Laboratory. (*USCO Mag.*, 1952, **3**, June, 9–11, 18). The redesigned laboratory at the Klip works of the Union Steel Corp. is described with illustrations. The various analytical operations carried out are given.—B. G. B.

MISCELLANEOUS

Science in Industry.—Sir James French. (*Engineer*, 1952, **194**, Sept. 19, 374–375). Since the war there has been an outcry, more among pure scientists than among engineers, that scientific discoveries are not sufficiently rapidly exploited in this country and that there is a great need of more technologists. Sir James French shows that the outcry is not a new one, but that, within the last forty years, a greater appreciation has grown amongst pure scientists of the merits of, and necessity for, research within industry by individual firms.

Industrial Research and Development. F. G. Pasotti. (*J. Junior Inst. Eng.*, 1952, **63**, Oct., 25–31). The purpose and work of development departments and research departments in industry are discussed.—B. G. B.

The Liége Section of the National Centre for Metallurgical Research. P. Coheur. (*Rev. Univ. Mines*, 1952, Series 9, **8**, Sept., 343–354). The organization of the section is explained and the close contact maintained with a large number of firms is stressed. Details are given of the varied research work carried out in the laboratories. A large part of the work is devoted to experiments, on a works scale, concerning the manufacture of basic-Bessemer steel ; brief details of the various experiments carried out are given. (44 references).

Cultural and Professional Aspects of Metallurgical Training. H. O'Neill. (*Bull. Inst. Met.*, 1952, **3**, Sept., 12–21 : *Engineering*, 1952, **174**, July 18, 94–95 : *Iron Coal Trades Rev.*, 1952, **165**, July 4, 27–30). Reference is made to the liberal aspects of training and welfare for metallurgical and other apprentices under the Guild system. Arguments are advanced for improving teaching of the philosophy and principles of science, and the wider cultural aspects of education, in University science courses. Moral aspects of the establishment of high standards of professional conduct and mutual assistance are discussed. (14 references).—K. E. J.

Administrative Control at Trostre : Production, Scheduling and Shipping. E. Williams. (*Iron Coal Trades Rev., Special Issue*, 1952, **165**, Oct. 1, 43–45, 74). A description is given of the administrative control of the scheduling, progressing, and

despatchment of orders at the Trostre Works of the Steel Company of Wales Ltd.—G. F.

Eye Accidents. W. Kölz. (*Roll'schen Eisenwerke Werkzeit*, 1952, **23,** Sept. 138–140). The dangers of eye injuries from splinters, acids, liquors, methyl violet, and exposure to heat and light rays are discussed. Details of preventive and curative measures are given.—E. C. S.

New Treatment of Silicosis. E. Rosenthal. (*Colliery Guardian*, 1952, **185,** Oct. 16, 461–464). A new method of treating silicosis in Germany is described. Minute droplets of calcium sodium salt, negatively charged, are kept in suspension in an electric field, and are inhaled by the patients.

Personnel Engagement at Trostre Works. D. O. Elliott. (*Iron Coal Trades Rev., Special Issue*, 1952, **165,** Oct. 1, 46–47). The author describes the system of recruitment and engagement of personnel at the Trostre Works of the Steel Company of Wales Ltd. Staff were mainly absorbed from the Company's existing works and from Richard Thomas and Baldwins, Ltd.

Personnel Services—Trostre. T. G. Watkins. (*Iron Coal Trades Rev., Special Issue*, 1952, **165,** Oct. 1, 48–49). A brief description is given of the Personnel Services Department and the amenities provided for employees at the Trostre Works of the Steel Company of Wales Ltd.—G. F.

The Engineer and Safety. F. M. Panzetta. (*J. Junior Inst. of Eng.*, 1952, **63,** Oct., 11–24). The author discusses safety precautions in factories.—B. G. B.

Medical Welfare. W. E. Moreton. (*Iron Coal Trades Rev., Special Issue*, 1952, **165,** Oct. 1, 50–52, 59). The organization and administration of the medical services at the Trostre Works of the Steel Company of Wales Ltd. are described.

Lighting Design, Planning and Layout. P. H. Billington. (*Mech. World*, 1952, **132,** Oct., 178–182). The choice of light source to suit particular situations and classes of work is discussed, and the spacing and location of fittings of various types examined. The recent developments in colour harmony are explained.—D. H.

Atmospheric Pollution. L. C. McCabe. (*Indust. Eng. Chem.*, 1952, **44,** Sept., 125A–128A). A recent investigation of smog at Los Angeles covers methods of analysis of pollutants, smog properties, emission, reduction in level, and effects on vegetation. Methods for collection and analysis are described for aldehydes, hydrocarbons, organic acids, ozone, total oxidants, oxides of nitrogen, hydro-peroxides, sulphur dioxide, carbon dioxide, aerosols, etc. A relative pollution index has been devised. A theory of smog formation is put forward.—K. E. J.

Water Supply at Port Talbot, Margam and Abbey Works. L. Griffiths. (*Iron Coal Trades Rev., Special Issue*, 1952, **165,** Oct. 1, 111–115). The author discusses the supply and use of domestic drinking water, fresh water, and dock water at the Port Talbot, Margam, and Abbey Works of the Steel Company of Wales Ltd.—G. F.

The De-aeration of Boiler Feed Water. H. Chambers. (*Metro. Vick. Gaz.*, 1952, **24,** May, 178–182). The unit described consists of one or more de-aerating chambers, a vapour condenser and a de-aerated-water storage tank with float control gear—all of which are combined to form a compact installation. A typical unit can de-aereate 45,000 gall./hr., reducing the oxygen content from 0·15 to 0·01 mls./litre.

Solids Fluidization Applied to Lime Burning. F. S. White and E. L. Kinsella. (*Min. Eng.*, 1952, **4,** Sept., 903–906). The solids fluidization technique can be applied to limestone calcination, giving fuel economy, high quality lime, and the possibility of varying final properties. Operating details and results are presented from a full-scale plant in which –6 mesh limestone is dried and dedusted in a single bed reactor, and then calcined in a 5-stage reactor.—K. E. J.

Organizing Cost Reduction. T. C. Gary. (*Mech. Eng.*, 1952, **74,** July, 551–554). The author discusses the basic principles for reducing costs in industry and illustrates his theories with details of the organization of cost reduction by the engineering department of the E.I. du Pont de Nemours and Co.—D. H.

Mechanical Handling in Industry and Its Effects on Costs. W. J. T. Dimmock. (*J. Inst. Prod. Eng.*, 1952, **31,** Feb., 90–119).

Estimating the Rating of Industrial Water and Steam Flow Meters. J. Tham and C. W. Terry. (*Instrument Engineer*, 1952, **1,** Oct., 28–29). The limited flow range of meters operating on the differential-pressure principle necessitates careful estimation of maximum and minimum flow conditions if meters are to be accurate and reliable. Factors influencing the choice of suitable maximum ratings of industrial water and steam flow meters are discussed.—J. C. B.

Distribution of Velocity and Mean Velocity in a Cylindrical Pipe. A. Mondiez. (*Chaleur et Ind.*, 1952, **33,** May, 151–152). This is a short contribution to the calculation of flow in pipes.—T. E. D.

Fluid Flow Through Orifices. R. J. Novotny. (*Product Eng.*, 1952, **23,** Apr., 213–217). Two nomographs are given, the first for determining the factors involved in the flow of fluids through orifices and the second for determining the effect of production tolerances on pressure drop across orifices. Examples of their use are detailed.—A. M. F.

Conductivity Type Heat Flow Meters. R. T. Fowler. (*Instrument Practice*, 1952, **6,** June, 527–533). A new heat flow meter is described. It uses the principle of heat conduction along a uniform rod, surrounded by a guard ring, with one end maintained at a constant temperature by cooling water. Two thermocouples, connected differentially, are inserted in the rod to determine the temperature difference over a fixed distance. The instrument can be used, without calibration, to measure radiant heat flows from sources at 800–1600° C. with an accuracy of 2%.—R. A. R.

Modern Feedwater Treatment. (*Eng. Boiler House Rev.*, **67,** Mar., 68–73). The author describes the production of demineralized water by means of hydrogen-ion exchange and acid adsorption materials. Using these in two columns in series, followed if necessary by a degassing tower and silica removal unit, water containing less than 5 p.p.m. total dissolved solids can be obtained. Using the latest 'mixed-bed' technique the whole process can be achieved in one column, and extremely pure water obtained.—D. L. C. P.

Safety. V. R. Croswell. (*Indust. Eng. Chem.*, 1952, **44,** July, 97A–100A). Some aspects of safety regulations, standards, and policy in chemical and other works in the U.S.A., Britain, and the Continent are discussed. Training and appeals to reason accomplish more for safety among the conservative British workers than do the contests, awards, and campaigns which have been successful in the U.S.A.—K. E. J.

Impaction of Dust and Smoke Particles on Surface and Body Collectors. W. E. Ranz and J. B. Wong. (*Indust. Eng. Chem.*, 1952, **44,** Jun., 1371–1381). Mechanisms of collection of dust and smoke particles are examined from the viewpoint of aerosol impaction on flat, cylindrical, and spherical collectors. An approximation method, involving dimensionless parameters, is proposed for determining impaction efficiency in any situation.—K. E. J.

Safety Analysis—An Aid to on-the-Job Instruction. A. R. Sauer. (*Iron Steel Eng.*, 1952, **29,** Aug., 133–136). The author makes an analytical approach to the accident prevention problem, involving the study of each job step by step, listing the potential hazards involved, and making recommendations for the recognition, elimination, or reduction of each hazard.—M. D. J. B.

Wire Guards. D. R. Woodley. (*Wire Production*, 1952, **1,** Sept., 18–22). The author deals with window and machinery guards designed to prevent accidents and vandalism. Design of guards, both technical and artistic, method of construction, and choice of wire mesh are discussed.—J. G. W.

Industrial Accidents. H. R. Payne. (*J. Roy. Soc. Arts*, 1952, **100,** Sept. 19, 740–752). The author shows that by wider adoption of methods which have been well proved in a great variety of industries, accidents could probably be reduced by 50%.—J. C. B.

Production Management Techniques. H. E. Kearsey. (*Metal Ind.*, 1952, **80,** May 16, 399–401; May 30, 439–411; June 6, 459–461; June 13, 487–489). The author discusses the attainment of higher productivity in the light of the reports of the several Anglo-American Productivity Teams. The attitude of labour, cost consciousness, pre-production planning, and the effectiveness of management are discussed in Part I. Part II is devoted to methods of study. The development of work values from these studies is discussed. In Part III, the application of incentive bonus schemes is dealt with, low-efficiency, 'pro rata,' and high-efficiency schemes being described. Standard costing is the subject of the final part.—P. M. C.

Management's Responsibility in Training. J. D. Judge. (*Amer. Foundryman*, 1952, **22,** July, 47–50). The author emphasizes the need for training semi-skilled and skilled

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workers and potential managers, bearing in mind also the need to develop each individual as a citizen.—E. T. L.

The Development and Training of Personnel. L. J. Westhaver. (*Amer. Iron Steel Inst. Regional Tech. Meetings*, 1950, 169–174). The author discusses the scheme adopted by the Geneva Steel Company for the selection, general training, and development of personnel.—G. F.

Industry and College Can Co-operate! W. T. Nichols. (*Chem. Eng. Prog.*, 1952, **48,** Apr., 201–206). The results of a survey, to determine how training courses and industry can co-operate, are outlined. The St. Louis Plan, promoted by the St. Louis section of the American Institute of Chemical Engineers is described.—T. E. D.

Some Impressions of Metallurgical Education in the U.S.A. K. M. Entwistle. (*Metallurgia*, 1952, **45,** Jan., 7–11). The O.E.E.C. Mission visited a number of educational institutions, and the author gives his personal impressions and discusses the differences in British and American methods. One difference noted was the inclusion of non-metallurgical subjects (English, foreign languages, economics) and engineering subjects in metallurgy courses.—B. G. B.

Iron and Steel. O. W. Ellis. (Symposium on Metallugical Research in Canada, 1951, Apr. ; *Canad. Min. Met. Bull.*, 1952, **45,** Apr., 218–219). Research on iron and steel going on at the universities and research organizations, and within the industry itself is briefly outlined.—T. E. D.

The Future of Industrial Research and Development in Great Britain. R. S. Edwards, (*Research*, 1952, **5,** June, 247–255). The position in Great Britain today is compared with that in the United States and Switzerland. Five points are made to improve Britain's prospects : (1) To hold our position in basic science ; (2) to increase the supply of qualified manpower to industry ; (3) the introduction of sponsored research to help as technological competition increases ; (4) capital accumulation burden for research must be reduced ; and (5) the pertinacity and initiative in industrial development should be rewarded by social esteem.—J. C. B.

Research at the Universities. B. Chalmers. (Symposium on Metallurgical Research in Canada), 1951, Apr. ; *Canad. Min. Met. Bull.*, 1952, **45,** Apr., 220–221). Fundamental research, of interest to the metallurgical industry, that is going on at the Universities of British Columbia and Toronto, and the Royal Military College at Kingston, is briefly discussed.

Alnico Recovery Process Salvages Valuable Nickel, Cobalt. A. H. Sherman and M. Pesses. (*Iron Age*, 1952, **170,** July 3, 115–119). This process recovers particles from refractory slags, grindings, and abrasive powders. Cleaned grindings and crushings are magnetically sorted before reblending in the correct proportions and aluminium is added as a deoxidizer. Silica and sodium silicate are added to produce a marketable slag. Before pouring, iron oxide is added to remove excess carbon and silicon.—A.M.F.

Physico-Chemical Studies on Dusts. III—Factors Affecting the Wetting of Airborne Silica Dusts by Aqueous Sprays. W. M. Cumming, W. H. S. Massie, and P. D. Ritchie. (*J. Appl. Chem.*, 1952, **2,** July, 413–422).

Atmospheric Pollution. L. C. McCabe. (*Indust. Eng. Chem.*, 1952, **44,** Aug., 105A–107A). Experimental work with a pilot fog-filter type scrubber has established its usefulness for absorption of fatty-acid odours, hydrogen sulphide, and sulphur dioxide from air streams. Full-scale installations have been made to treat hydrogen sulphide and fatty acids, and are planned for sulphur dioxide and trioxide vapours. Apparatus is also described for condensing blanketing steam, and thereby reducing load in fume-disposal equipment for an asphalt oxidizing plant.—K. E. J.

Heat Relief on Hot Summer Days in Hot Industrial Operations. W. Schweisheimer. (*Indust. Heating*, 1952, **19,** July, 1222–1228). Among the successful methods for combating high temperatures are the use of salt tablets, circulating fans, exhaust ventilation, and insulation from radiant heat. The effect of humidity and absenteeism due to sunburn are considered.—B. G. B.

The Measurement of Fluidization Quality. W. W. Shuster and P. Kisliak. (*Chem. Eng. Prog.*, 1952, **48,** Sept., 455–458). An instrument is described for evaluating the quality of fluidization based on the measurement of the magnitude and frequency of pressure fluctuations across a section of a fluidized bed. This quality meter was tested under varying conditions of bead size, initial bed height, column diameter, and bed expansion.—T. E. D.

Thoughts on Symbols for Steels and their Standardization in Italy. G. Z. Zazinovich. (*Rev. Mét.*, 1952, **49,** Aug., 573–584). The system in use in Italy is described. Simple or carbon steels are subdivided according to use into ordinary and special groups. Alloy steels are grouped into those with additions greater or less than 5%. Further subdivisions and the symbols used to indicate both type and composition are described.—A. G.

Systematic Identification Markings of Metallic Materials. E. Siebel and F. Fischer. (*Metall u. Erz*, 1942, **39,** Dec., No. 24, 442–445 ; 1944, **41,** July, No. 13–14, 155–157). A system of marking is described, which indicates concisely, by a few symbols and a simple code, the method of manufacture, composition, and some physical properties of metals. Examples are given of its application to ferrous and non-ferrous materials.

BOOK NOTICES

CORDERO, H. S. "*Iron and Steel Works of the World.*" 8vo, pp. 651. London, 1952 : Quin Press, Ltd. (Price 50*s.*)

This publication aims to describe all the important producers of pig iron, raw steel and rolled steel ; further into the fabricating processes it does not go, save where the producers of semi-finished products themselves operate foundries and forging shops. Nor does it include re-rollers. Within its limits, this task is carried out extremely well ; information on Europe and on North America (identical in this volume, with " Directory of Iron and Steel Works in the United States and Canada, 1951 " published by the American Iron and Steel Institute) may be, perhaps, taken for granted, but the details of operations in other continents are not to be found so easily elsewhere. Where else could one discover, for example, that a steel industry exists in the Philippines, based at the present on scrap, but later to use charcoal pig of local manufacture, or that an integrated works of medium size, to produce 160,000 tons of basic-Bessemer steel annually, is being erected in Colombia ?

Other evidence of the scope of the volume is to be found in the excellent section, with maps, on plants in the U.S.S.R., and in the listing of the new German steel concerns to be assembled from the remnants of the former great combines. In all, the book will certainly be of value and interest to everyone concerned in the steel industry.—J. P. S.

INTERNATIONAL LABOUR ORGANISATION, IRON AND STEEL COMMITTEE. Report I. "*General Report.*" 8vo, pp. iv + 106. Report II. "*Vocational Training and Promotion in the Iron and Steel Industry.*" 8vo, pp. iv + 94. Report III. "*Welfare Services in the Iron and Steel Industry.*" 8vo, pp. iv + 80. Fourth Session, Geneva, 1952. Geneva, 1952 : International Labour Office.

The agenda for the fourth session of the Iron and Steel Committee held in June, 1952, was prepared by the International Labour Organisation (I.L.O.) during 1950, and these three Reports were issued for consideration.

Report I.—General Report.—This is a most interesting and timely review of fluctuations since 1949 of supply and demand in the iron and steel markets of the world. There is reference to the quick changes from a seller's to a buyer's and back to a seller's market, and to the present critical position—particularly in European countries—arising from inadequate supplies of coal, coke, iron ore, and steel scrap. Production results for 32 countries are followed by an analysis of the modernization and development plans being carried out in the iron and steel industries of those countries.

The third chapter of this book, on world iron and steel affairs, deals with labour questions—employment, wages, hours of work, and industrial relations. This is followed by reviews explaining the introduction of co-determination in the German iron and steel industry ; the setting up of a Metals Council in Belgium ; the principal social provisions of the treaty for the European Coal and Steel Community ; and finally the methods of international co-operation in the industry.

The very radical developments in Germany and Belgium concerning industrial co-operation or joint consultation are of special interest, because they may have quite a wide influence on the Continent through the Schuman Treaty Organization.

Report II.—Vocational Training and Promotion in the Iron and Steel Industry.—In one chapter there is the apology that it " does not claim to be either perfect or complete." Similar or even stronger comments could have been made about this particular Report, but, because of inaccuracies, a " Report II (Supplement) " was issued by the I.L.O. towards the end of 1952. It is important that the reader should see the new text, because the material in it now conveys a correct impression of training in the British Steel industry.

Notwithstanding this criticism, the original Report II endeavours to present world-wide problems peculiar to the greater development of planning of vocational training and to the raising of the level of theoretical education of workers and supervisors in the production trades of the iron and steel industry. There is a tendency for the Report to be over-ambitious. It suggests national classifications and descriptions of skills and occupations to provide bases for analysing the technical and physical qualifications for any given job, and for drawing up standard training curricula. In this latter connection it asks that consideration be given to practical, methodical, and comprehensive vocational training ; theoretical training, including courses of general education ; and physical training of entrants to skilled occupations.

The second part of this Report discusses the differing policies of up-grading and promoting workers in the industry. In some countries upgrading of workers from one grade to the next is based on seniority—and there are also varying definitions of seniority. In other countries up-grading is decided solely by the managements, and there is a tendency to provide facilities for what—in Britain—might be regarded as accelerated up-grading. Promotion of workers to supervisory positions is usually a prerogative of management, but the question is raised as to whether or not a worker who applies for admission to a supervisory training course and has been rejected, should have the right of appeal to another body including workers' representatives. There is also reference to merit rating of employees and to the lead in this field being held by the United States.

The Supplement previously referred to supersedes the two sections of the original report which deal with the vocational training in the British iron and steel industry. The supplement is a factual summary of the position since the establishment in 1947 of the British Iron and Steel Federation Training Committee.

Report III.—Welfare Services in the Iron and Steel Industry.—This Report surveys those matters which affect the general well-being and morale of iron and steel workers. Of necessity, this brief summary cannot go into detail, but it gives many interesting examples of legislative and industrial initiative in the development of employee services, and it presents a broad picture of trends in the provision of welfare in the plants, outlines the extent to which the steel industry in many countries enters into the field of general social welfare, and compares the provisions for the participation by the workpeople in the organization and management of welfare services.

The Report would reveal little new to a British employer as regards the various kinds of provisions generally made, but it provides useful comparisons, and it lists many welfare items for discussion at international level.

These three Reports have been discussed this year by the I.L.O. Iron and Steel Committee. They present most useful international comparisons and raise many of the problems existing in the labour field of the iron and steel industries. Britain is at the half-way stage in her plant development plans, and this may be a useful time to study, within individual organizations and at national level, any desirable revisions to employment policies.—J. WADSWORTH

MORLEY, ARTHUR. " *Strength of Materials.*" Tenth Edition. 8vo, pp. x + 583. Illustrated. London, New York, Toronto, 1952 : Longmans, Green and Co. (Price 25*s.*)

This is, of course, a classical textbook of high repute which has been in use for approximately 45 years, and one notes with interest that, with the present edition, 97,000 copies have been issued. The previous edition was published in 1940. In the present edition, " Many small changes and additions have been made, particularly in relation to helical springs, strain gauges, and strain analysis."

It is unnecessary to comment on the style and quality of the general matter which have proved themselves beyond words. However, a metallurgist naturally turns to those chapters containing matter of direct interest to him. The relevant material here has, as is common with such engineering texts, a somewhat old-fashioned aspect, although it is soundly written, apart from occasional statements such as, " Drawn copper, brass and bronzes are softened by the process of quenching or quick cooling from a high temperature..." (p. 62). Admittedly, metallurgical discussion plays only a supporting part, but it is a pity that it has not been possible to include some details of more recent knowledge in the sections on creep, fatigue, and effect of temperature on mechanical properties.

A. R. BAILEY

UNITED NATIONS. " *United Nations Scientific Conference on the Conservation and Utilization of Resources.*" Volume II. " *Mineral Resources.*" 8vo, pp. 303. New York, 1951 : United Nations, Department of Economic Affairs ; London : H.M. Stationery Office. (Price $3)

This publication presents a large number of papers on the measurement, discovery, and conservation of minerals, given at a conference at Lake Success in August and September, 1949. Not all will be of interest to ferrous metallurgists, but worthy of note are " Mineral Supplies and their Measurement (Iron-Manganese)," by A. Legendre (pp. 4–11), " Outlook for Mineral Discovery in Great Britain," by W. F. P. McLintock (pp. 44–47), " Conservation in Production of Iron and Steel," by C. Williams (pp. 172–176), " The Costs of Corrosion and its Control," by W. H. J. Vernon (pp. 218–222) and " Possibilities and Limitation in the Substitution of Light Metals for Steel and Copper," by J. Matter, (pp. 254–256). It is fairly clear that the highest grades, and most easily accessible deposits, of many industrial minerals are nearing exhaustion ; it is comforting to find that there is a growing endeavour to meet the results of this process by a variety of means.—J. P. S.

NEW PUBLICATIONS

AITCHISON, LESLIE, and WILLIAM I. PUMPHREY. " *Engineering Steels.*" A Study of the Properties of Steels and the Principles Governing their Selection for Engineering Applications. 8vo, pp. 1000 (approx.). Illustrated. London, 1952 : MacDonald and Evans. (Price £5 5*s.*)

AMERICAN SOCIETY OF MECHANICAL ENGINEERS ; SOCIETY FOR THE ADVANCEMENT OF MANAGEMENT. " *Techniques of Plant Maintenance*—1952." Third Plant Maintenance Conference, January, 1952. La. 8vo, pp. 182. New York, 1952 : Clapp and Poliak, Inc. (Price $6)

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8vo, pp. 20. Illustrated. London, 1952 : The Institution. (Price 4*s.*)

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COLEMAN, H. S. "*Laboratory Design.*" La. 8vo, pp. 393. Illustrated. New York, 1952 : Reinhold Publishing Corp. (Price $15)

DELBART, GEORGES. "*Les Laboratoires de l'Institut de Recherches de la Sidérurgie Française.*" La. 8vo, pp. 48. Illustrated. St. Germain-en-Laye (S. et O.), 1952 : The Institute.

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